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(54) **STEEL SHEET AND METHOD FOR PRODUCING SAME**

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(57) **ABSTRACT**

A steel sheet of the present invention is a steel sheet having a predetermined chemical composition and containing at least ferrite, residual austenite, and/or martensite in a micro-structure, and furthermore, is a steel sheet in which, in a plane parallel to a rolled surface, an average distance between centers of high Mn regions adjacent to each other is 1.00 mm or less, a density D_A of the high Mn regions at a sheet width center portion and a density D_B of the high Mn regions at a ¼ position from a sheet width end portion satisfy a relationship of $0.77 \leq D_A/D_B \leq 1.30$, a ratio of an average hardness of the high Mn regions to an average hardness of the low Mn regions is 1.1 to 2.0, and a difference between an average of a top 5% and an average of a bottom 5% of Mn contents in the low Mn regions is 0.3% or more.

9 Claims, No Drawings

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STEEL SHEET AND METHOD FOR PRODUCING SAME

TECHNICAL FIELD OF THE INVENTION

The present invention relates to a steel sheet and a method for producing the same. Priority is claimed on Japanese Patent Application No. 2019-075693, filed Apr. 11, 2019, the content of which is incorporated herein by reference.

BACKGROUND ART

In recent years, awareness of environmental issues has increased, and in the automobile industry, it is important to reduce the weight of a vehicle body in order to improve fuel efficiency. In order to achieve both a reduction in the weight and an improvement in the safety of the vehicle body, the use of a high strength material (high strength steel) is being studied. However, the higher the strength of the steel, the more difficult it is to perform press forming, and there is a problem that shape-fixability decreases (the shape of the steel is likely to collapse due to springback). In addition, the higher the strength, the lower the ductility, so that fracture is likely to occur during press forming. Furthermore, even if steel sheets of the same coil are pressed by the same method, the shapes thereof become slightly different. In other words, there is also a problem that the higher the strength, the lower the dimensional precision.

As a result of examinations by the present inventors, it was found that springback is likely to occur due to some portions where the steel does not yield. Therefore, it was found that if it is possible to lower the yield stress while increasing the maximum strength (tensile strength) of the steel, the shape-fixability can be easily improved. On the other hand, it was found that in order to increase the dimensional precision, a uniform metallographic structure may be formed throughout a steel sheet and changes depending on the location of the steel sheet may be reduced. Furthermore, it was found that although it is natural that the dimensional precision is increased when changes in strength and ductility are small, since the shape of a steel is formed through work hardening during press forming, the dimensional precision can be further increased when the difference in work hardening due to the location can be reduced.

As a high strength material applied to the vehicle body of a vehicle, composite structure steels such as dual phase (DP) steels described in Patent Documents 1 and 2 and transformation induced plasticity (TRIP) steels described in Patent Documents 3 and 4 are known.

DP steels are increased in the strength by allowing a full hard structure to be present in the steel. Furthermore, DP steels are characterized in that the amount of work hardening is increased even in a high strain region in order to increase ductility. In DP steels, the presence of martensite in the steel also allows the presence of moving dislocations in the periphery and reduces the yield stress.

In addition, the TRIP steels are further increased in the amount of work hardening through the strain-induced transformation of residual austenite so as to be less likely to be fractured, thereby increasing ductility.

These steel sheets allow the full hard structure to be dispersed and are increased in the amount of work hardening. However, when the amount of the full hard structure changes slightly, a work hardening method changes, although the yield stress and the tensile strength do not change significantly. When the amount of work hardening during forming changes, the amount of change in the shape

of the steel during forming also changes depending on the location or by the sheet itself, resulting in poor dimensional precision.

Patent Document 5 describes a method in which a method for applying water to a slab during casting, particularly until solidification, is set or the amount of water is set to a specific range to control the cooling rate, thereby controlling the segregation of Mn and P. An object of Patent Document 5 is to control the irregularities of the surface after working, and it is also possible to reduce the difference in work hardening depending on the location. However, Patent Document 5 targets a steel sheet having a strength of less than 590 MPa, and the C content or the Mn content thereof, particularly the C content is small. As a result of examinations by the present inventors, it was found that the segregation cannot be sufficiently controlled only by controlling the cooling rate until solidification as described in Patent Document 5 in a steel having a C content or Mn content required for a high strength steel sheet of 590 MPa or more, and the difference in work hardening depending on the location cannot be reduced.

As in Patent Document 5, an object of Patent Document 6 is to control segregation during casting, and a method in which the amount of water applied to a slab during casting is adjusted to control the cooling rate, thereby controlling Mn segregation. However, as described in Patent Document 6, only the surface portion can be controlled during solidification. In Patent Document 6, although the above method is sufficient since the object is to improve bendability in which segregation of the surface is important, in the case of improving press formability including tensile properties, a concentration distribution depending on the location has to be reduced by controlling segregation not only on the surface of the sheet but also at a position closer to the center portion. However, the technique of Patent Document 6 cannot control the segregation of the center portion.

That is, in the related art, regarding a high strength steel sheet of 590 MPa or more, a steel sheet having excellent shape-fixability and dimensional precision after pressing has not been proposed.

PRIOR ART DOCUMENT

Patent Document

[Patent Document 1] Japanese Patent No. 5305149

[Patent Document 2] Japanese Patent No. 4730056

[Patent Document 3] Japanese Unexamined Patent Application, First Publication No. S61-157625

[Patent Document 4] Japanese Unexamined Patent Application, First Publication No. 2007-063604

[Patent Document 5] Japanese Unexamined Patent Application, First Publication No. 2018-145525

[Patent Document 6] Japanese Unexamined Patent Application, First Publication No. 2012-219341

DISCLOSURE OF THE INVENTION

Problems to be Solved by the Invention

In view of the current status of the related art, an object of the present invention is to provide, as a high strength steel sheet having a tensile strength of 590 MPa or more, which is suitable as a steel sheet for a vehicle subjected to press working, a steel sheet having sufficient workability and being excellent in shape-fixability and dimensional precision after pressing, and a method for producing the same.

Means for Solving the Problem

The present inventors intensively studied a method for solving the above problems and obtained the following findings.

It was found that by controlling the chemical composition and the structure of a steel sheet, and controlling the distributed state of Mn and hardness in a plane (¼ plane) parallel to a rolling direction at a ¼ thickness position in a sheet thickness direction from the surface of the steel sheet, a change in the amount of work hardening depending on the location can be reduced, whereby it is possible to produce a steel sheet having a low yield stress and excellent formability.

In addition, it was found that in order to obtain the above distribution, it is effective to apply a pressure to a slab during cooling of the slab while controlling an average cooling rate in a predetermined temperature range, to control a heating rate during heating for hot rolling, and then perform hot rolling, cooling, coiling, pickling, cold rolling, annealing, and cooling including retention in the middle under predetermined conditions.

The present invention has been made based on the above findings, and the gist thereof is as follows.

(1) A steel sheet including, as a chemical composition, by mass %: C: 0.040% to 0.400%; Si: 0.01% to 2.50%; Mn: 0.10% to 4.00%; Al: 0.010% to 1.500%; P: 0.001% to 0.100%; S: 0.0005% to 0.0100%; N: 0.0005% to 0.0100%; Ti: 0% to 0.200%; Mo: 0% to 0.300%; Nb: 0% to 0.200%; Cr: 0% to 4.00%; B: 0% to 0.0050%; V: 0% to 0.300%; Ni: 0% to 4.00%; Cu: 0% to 4.00%; W: 0% to 2.00%; Ca: 0% to 0.0100%; Ce: 0% to 0.0100%; Mg: 0% to 0.0100%; Zr: 0% to 0.0100%; La: 0% to 0.0100%; REM other than Ce and La: 0% to 0.0100%; Sn: 0% to 1.000%; Sb: 0% to 0.200%; and a remainder: Fe and impurities, in which a microstructure in a range from a ⅛ thickness position in a sheet thickness direction from a surface of the steel sheet to a ⅜ thickness position in the sheet thickness direction from the surface includes, by area fraction, ferrite: 10% to 97%, residual austenite and martensite: 3% to 90%, bainite: 0% to 87%, and pearlite: 0% to 10%, in a plane parallel to a rolling direction at a ¼ thickness position in the sheet thickness direction from the surface, when a maximum value of Mn contents in a measurement range is indicated as Mnmax, an average value of the Mn contents is indicated as Mnave, regions where the Mn content is (Mnave+Mnmax)/2 or more are indicated as high Mn regions, and the other regions are indicated as low Mn regions, an average distance between centers of the high Mn regions adjacent to each other is 1.00 mm or less, a density D_A of the high Mn regions at a sheet width center portion and a density D_B of the high Mn regions at a ¼ width position from a sheet width end portion satisfy Expression (1), a ratio of an average hardness of the high Mn regions to an average hardness of the low Mn regions is 1.1 to 2.0, and a difference between an average of a top 5% and an average of a bottom 5% of the Mn contents at measurement points in the low Mn regions is 0.3 mass % or more.

$$0.77 \leq D_A/D_B \leq 1.30 \quad \text{Expression (1)}$$

(2) The steel sheet according to (1), in which a hot-dip galvanized layer is formed on the surface.

(3) The steel sheet according to (2), in which the hot-dip galvanized layer is a hot-dip galvannealed layer.

(4) A method for producing the steel sheet according to (1), including: a casting step of producing a slab by melting a steel having the chemical composition according to (1), casting the melted steel to produce a slab, and cooling the slab at a temperature of 950° C. to 550° C. while applying a pressure of 10 N/cm^2 or more to the slab in a sheet thickness direction so that an average cooling rate is 100° C./h or faster; a heating step of heating the slab to a temperature range of 1100° C. to 1280° C. after cooling the slab to room temperature or before cooling the slab to room temperature so that an average heating rate in a temperature range of 650° C. to 850° C. is 50° C./min or slower; a hot rolling step of hot-rolling the slab after the heating step in a temperature range of 1050° C. or higher at a cumulative rolling reduction of 35% or more to obtain a hot-rolled steel sheet; a cooling step of cooling the hot-rolled steel sheet to 650° C. or lower, the cooling being started within three seconds after completion of the hot rolling step, so that an average cooling rate to 700° C. is 20° C./s or faster; a coiling step of coiling the hot-rolled steel sheet after the cooling step in a temperature range of 300° C. to 650° C.; a pickling step of performing pickling on the hot-rolled steel sheet after the coiling step to obtain a pickled steel sheet; a cold rolling step of performing cold rolling on the pickled steel sheet to obtain a cold-rolled steel sheet; an annealing step of heating the cold-rolled steel sheet to an annealing temperature of Ac1° C. to 1000° C. at an average heating rate of 10.0° C./s or slower and performing holding at the annealing temperature for five seconds to 600 seconds; a post-annealing cooling step of cooling the cold-rolled steel sheet after the annealing step to a retention temperature of 150° C. to 550° C. at an average cooling rate of 1° C./s to 200° C./s; a retaining step of performing retention at the retention temperature for 15 seconds to 1000 seconds; and a final cooling step of cooling the cold-rolled steel sheet after the retaining step to room temperature.

(5) The method for producing the steel sheet according to (4), further including: a hot-dip galvanizing step of immersing the cold-rolled steel sheet in a molten zinc bath, between the retaining step and the final cooling step.

(6) The method for producing the steel sheet according to (5), further including: an alloying step of reheating the cold-rolled steel sheet to 470° C. to 550° C. and performing holding for 60 seconds or shorter, between the hot-dip galvanizing step and the final cooling step.

(7) The method for producing the steel sheet according to (4) to (6), further including: a leveling step of working the cold-rolled steel sheet using a leveler, between the cold rolling step and the annealing step.

Effects of the Invention

According to the present invention, it is possible to provide a high strength steel sheet having a tensile strength of 590 MPa or more, sufficient workability, and shape-fixability and high dimensional precision after pressing, and a method for producing the same. This high strength steel sheet is suitable as a steel sheet for a vehicle subjected to press working.

EMBODIMENTS OF THE INVENTION

A steel sheet according to an embodiment of the present invention (a steel sheet according to the present embodiment) contains,

(i) as a chemical composition, by mass %, C: 0.040% to 0.400%, Si: 0.01% to 2.50%, Mn: 0.10% to 4.00%, Al:

0.010% to 1.500%, P: 0.001% to 0.100%, S: 0.0005% to 0.0100%, and N: 0.0005% to 0.0100%, optionally contains Ti, Mo, Nb, Cr, B, V, Ni, Cu, W, Ca, Ce, Mg, Zr, La, REM other than Ce and La, Sn, and Sb, and contains a remainder consisting of Fe and impurities,

(ii) in which, a microstructure at a ¼ thickness includes, by area fraction, ferrite: 10% to 97%, residual austenite and martensite: 3% to 90%, bainite: 0% to 87%, pearlite: 0% to 10%,

in a plane parallel to a rolled surface at the ¼ thickness, when a maximum value of Mn contents in a measurement range is indicated as Mnmax, an average value of the Mn contents is indicated as Mnave, regions where the Mn content is (Mnave+Mnmax)/2 or more are indicated as high Mn regions, and the other regions are indicated as low Mn regions,

(iii) an average distance between centers of the high Mn regions adjacent to each other is 1.00 mm or less,

(iv) a density D_A of the high Mn regions at a sheet width center portion and a density D_B of the high Mn regions at a ¼ width position from a sheet width end portion satisfy $0.77 \leq D_A/D_B \leq 1.30$,

(v) a ratio of an average hardness of the high Mn regions to an average hardness of the low Mn regions is 1.1 to 2.0, and

(vi) a difference between an average of a top 5% and an average of a bottom 5% of the Mn contents at measurement points in the low Mn regions is 0.3 mass % or more.

Hereinafter, the steel sheet of the present embodiment and a method for producing the steel sheet according to the present embodiment will be sequentially described.

First, the reason for limiting the chemical composition of the steel sheet according to the present embodiment will be described. Hereinafter, % regarding each element in the chemical composition means mass %.

C: 0.040% to 0.400%

C is an element that contributes to an increase in the fraction of martensite and an improvement in the strength of martensite. When the C content is less than 0.040%, it is difficult to obtain the tensile strength (590 MPa or more) required for a high strength steel sheet. Therefore, the C content is set to 0.040% or more. The C content is preferably 0.050% or more.

On the other hand, when the C content exceeds 0.400%, point weldability deteriorates. Therefore, the C content is set to 0.400% or less. The C content is preferably 0.350% or less, and more preferably 0.300% or less.

Si: 0.01% to 2.50%

Si is an element that contributes to an improvement in tensile strength and fatigue strength without lowering ductility through solid solution strengthening. Si is also an element having a deoxidizing effect. When the Si content is less than 0.01%, the above effect cannot be sufficiently obtained. Therefore, the Si content is set to 0.01% or more. The Si content is preferably 0.03% or more.

On the other hand, when the Si content exceeds 2.50%, segregation of Mn is promoted, so that the difference in the amount of work hardening depending on the location of the steel sheet increases, or ductility and spot toughness decrease. Therefore, the Si content is set to 2.50% or less. The Si content is preferably 2.00% or less.

Mn: 0.10% to 4.00%

Mn is an element that contributes to the improvement in the strength of the steel by improving solid solution strengthening and hardenability. When the Mn content is less than 0.10%, the above effect cannot be sufficiently obtained. Therefore, the Mn content is set to 0.10% or more. The Mn content is preferably 0.30% or more, and more preferably 1.00% or more.

On the other hand, when the Mn content exceeds 4.00%, the weldability decreases, and there are cases where segregation expands and formability during pressing decreases, or there are cases where the steel sheet being produced is cracked during a production process. In addition, when the segregation of Mn increases, the difference in the amount of work hardening depending on the location of the steel sheet increases. Therefore, the Mn content is set to 4.00% or less. The Mn content is preferably 3.80% or less, and more preferably 3.00% or less.

Al: 0.010% to 1.500%

Al is an element necessary for deoxidation, and is also an element that contributes to an improvement in the formability by suppressing excessive generation of carbides. When the Al content is less than 0.010%, the above effect cannot be sufficiently obtained. Therefore, the Al content is set to 0.010% or more. The Al content is preferably 0.020% or more.

On the other hand, when the Al content exceeds 1.500%, the above effect is saturated, and the segregation of Mn is promoted, so that the difference in the amount of work hardening depending on the location of the steel sheet increases, or the ductility and spot toughness decrease. Therefore, the Al content is set to 1.500% or less. The Al content is preferably 1.000% or less.

P: 0.001% to 0.100%

P is an element that contributes to the improvement in the strength, and is an element that enhances corrosion resistance in the coexistence with Cu. When the P content is less than 0.001%, the effect cannot be sufficiently obtained. In addition, in order to cause the P content to be less than 0.001%, a steelmaking cost increases significantly. Therefore, the P content is set to 0.001% or more. From the viewpoint of the steelmaking cost, the P content is preferably 0.010% or more.

On the other hand, when the P content exceeds 0.100%, the weldability and formability decrease. Therefore, the P content is set to 0.100% or less. The P content is preferably 0.020% or less.

S: 0.0005% to 0.0100%

S is an element that forms a sulfide (MnS or the like) that is an origin of cracking and reduces hole expansibility and total elongation. Although the S content may be as small as possible, when the S content is reduced to less than 0.0005%, the steelmaking cost increases significantly. Therefore, the S content is set to 0.0005% or more. The S content is preferably 0.0010% or more.

On the other hand, when the S content exceeds 0.0100%, the formability significantly decreases. Therefore, the S content is set to 0.0100% or less. The S content is preferably 0.0060% or less.

N: 0.0005% to 0.0100%

N is an element that inhibits the workability. In addition, N is an element that forms a nitride (TiN and/or NbN) that decreases the formability in the coexistence with Ti and/or Nb and thus reduces the effective amount of Ti and/or Nb.

Although the N content may be as small as possible, when the N content is reduced to less than 0.0005%, the steelmaking cost increases significantly. Therefore, the N content is set to 0.0005% or more. The N content is preferably 0.0012% or more.

On the other hand, when the N content exceeds 0.0100%, the formability significantly decreases. Therefore, the N content is set to 0.0100% or less. The N content is preferably 0.0060% or less.

The steel sheet according to the present embodiment may contain the above elements, and the remainder consisting of Fe and impurities. However, for the purpose of further improving the properties, the steel sheet may include one or two or more selected from the group consisting of Ti: 0.200% or less, Mo: 0.300% or less, Nb: 0.200% or less, Cr: 4.00% or less, B: 0.0050% or less, V: 0.300% or less, Ni: 4.00% or less, Cu: 4.00% or less, W: 2.00% or less, Ca: 0.0100% or less, Ce: 0.0100% or less, Mg: 0.0100% or less, Zr: 0.0100% or less, La: 0.0100% or less, REM other than Ce and La: 0.0100% or less, Sn: 1.000% or less, and Sb: 0.200% or less. Since these elements do not have to be contained, the lower limit thereof is 0%.

Ti: 0% to 0.200%

Ti is an element that contributes to the formation of unrecrystallized ferrite by delaying recrystallization, and contributes to the improvement in the strength by forming carbides and/or nitrides.

When Ti is less than 0.010%, there are cases where the above effect of containing Ti is not sufficiently obtained. Therefore, Ti is preferably 0.010% or more.

On the other hand, when the Ti content exceeds 0.200%, the segregation of Mn is promoted, and the difference in the amount of work hardening depending on the location of the steel sheet increases. Therefore, the Ti content is set to 0.200% or less. The Ti content is preferably 0.100% or less, and more preferably 0.050% or less.

Mo: 0% to 0.300%

Mo is an element that enhances hardenability and contributes to the control of a martensite fraction. Mo is also an element that segregates to grain boundaries, suppresses zinc from infiltrating into the structure of a weld during welding, and contributes to the prevention of cracking during welding. In addition, Mo is an element that contributes to the suppression of the generation of pearlite during cooling in an annealing step.

When Mo is less than 0.010%, there are cases where the above effect of containing Mo is not sufficiently obtained. Therefore, the Mo content is preferably 0.010% or more. The Mo content is more preferably 0.040% or more.

On the other hand, when the Mo content exceeds 0.300%, the formability deteriorates. Therefore, the Mo content is set to 0.300% or less. The Mo content is preferably 0.250% or less.

Nb: 0% to 0.200%

Nb is an element that contributes to the formation of unrecrystallized ferrite by delaying recrystallization, and contributes to the improvement in the strength by forming carbides and/or nitrides. When the Nb content is less than 0.005%, there are cases where the above effect of containing Nb is not sufficiently obtained. Therefore, the Nb content is preferably 0.005% or more. The Nb content is more preferably 0.010% or more.

On the other hand, when the Nb content exceeds 0.200%, the segregation of Mn is promoted, and the difference in the amount of work hardening depending on the location of the steel sheet increases. Therefore, the Nb content is set to 0.200% or less. The Nb content is preferably 0.170% or less.

Cr: 0% to 4.00%

Cr is an element that contributes to the suppression of the generation of pearlite during cooling in an annealing step. When the Cr content is less than 0.01%, there are cases where the above effect of containing Cr is not sufficiently obtained. Therefore, the Cr content is preferably 0.01% or more. The Cr content is more preferably 0.05% or more.

On the other hand, when the Cr content exceeds 4.00%, the formability decreases. Therefore, the Cr content is set to 4.00% or less. The Cr content is preferably 1.50% or less.

B: 0% to 0.0050%

B is an element that enhances hardenability and contributes to the control of a martensite fraction. B is also an element that segregates to grain boundaries, suppresses zinc from infiltrating into the structure of a weld, and contributes to the prevention of cracking during welding. In addition, B is an element that contributes to the suppression of the generation of pearlite during cooling after annealing. Furthermore, B is also an element that contributes to an improvement in toughness through grain boundary strengthening during boundary segregation.

When the B content is less than 0.0002%, there are cases where the above effect is not sufficiently obtained. Therefore, the B content is preferably set to 0.0002% or more. The B content is more preferably 0.0010% or more.

On the other hand, when the B content exceeds 0.0050%, boride is generated and the toughness decreases. Therefore, the B content is set to 0.0050% or less. The B content is preferably 0.0025% or less.

V: 0% to 0.300%

V is an element that contributes to the improvement in the strength by precipitate strengthening, grain refinement strengthening by suppressing the growth of grains, and dislocation strengthening by suppressing recrystallization. When the V content is less than 0.001%, there are cases where the strength improving effect is not sufficiently obtained. Therefore, the V content is preferably 0.001% or more. The V content is more preferably 0.010% or more.

However, when the V content exceeds 0.300%, carbonitrides are excessively precipitated and the formability decreases. Therefore, the V content is set to 0.300% or less. The V content is preferably 0.150% or less.

Ni: 0% to 4.00%

Ni is an element that suppresses phase transformation at high temperatures and contributes to the improvement in the strength. When the Ni content is less than 0.01%, there are cases where the above effect is not sufficiently obtained. Therefore, the Ni content is preferably set to 0.01% or more. The Ni content is more preferably 0.10% or more.

On the other hand, when the Ni content exceeds 4.00%, the weldability decreases. Therefore, the Ni content is set to 4.00% or less. The Ni content is preferably 3.50% or less.

Cu: 0% to 4.00%

Cu is an element that exists as fine particles in steel and contributes to the improvement in the strength. When the Cu content is less than 0.01%, there are cases where the above effect is not sufficiently obtained. Therefore, the Cu content is preferably 0.01% or more. The Cu content is more preferably 0.10% or more.

On the other hand, when the Cu content exceeds 4.00%, the weldability decreases. Therefore, the Cu content is set to 4.00% or less. The Cu content is preferably 3.50% or less.

W: 0% to 2.00%

W is an element that suppresses phase transformation at high temperatures and contributes to the improvement of the strength of steel. When the W content is less than 0.01%, there are cases where the above effect is not sufficiently obtained. Therefore, the W content is preferably set to 0.01% or more. The W content is more preferably 0.10% or more.

On the other hand, when the W content exceeds 2.00%, hot workability decreases and productivity decreases. Therefore, the W content is set to 2.00% or less. The W content is preferably 1.20% or less.

Ca: 0% to 0.0100%

Ce: 0% to 0.0100%

Mg: 0% to 0.0100%

Zr: 0% to 0.0100%

La: 0% to 0.0100%

REM other than Ce and La: 0% to 0.0100%

Ca, Ce, Mg, Zr, La, and REM are elements that contribute to the improvement in the formability. When each of Ca, Ce, Mg, Zr, La, and REM other than Ce and La is less than 0.0001%, there are cases where the effect of containing the elements is not sufficiently obtained. Therefore, in a case where these elements are contained, the amount of each of the elements is preferably 0.0001% or more. More preferably, the amount of each of the elements is 0.0010% or more.

When the amount of each of Ca, Ce, Mg, Zr, La, and REM other than Ce and La exceeds 0.0100%, there is concern that the ductility may decrease. Therefore, the amount of any of the elements is set to 0.0100% or less. Preferably, the amount of any of the elements is 0.0070% or less.

REM is an abbreviation for Rare Earth Metal and here, refers to elements belonging to lanthanoid series excluding Ce and La, Sc, and Y. Since Ce and La exhibits the above effects compared to other elements belonging to lanthanoid series, Ce and La are excluded from REM in the steel sheet according to the present embodiment. REM is often contained in the form of mischmetal, but there are cases where elements of the lanthanoid series are contained in combination. Even if an element of the lanthanoid series is contained as an impurity, the property is not impaired.

Sn: 0% to 1.000%

Sn is an element that suppresses the coarsening of the structure and contributes to the improvement in the strength. When Sn is less than 0.001%, the above effect of containing Sn is not sufficiently obtained. Therefore, the Sn content is preferably 0.001% or more. The Sn content is more preferably 0.010% or more.

On the other hand, when the Sn content exceeds 1.000%, the steel sheet may be excessively embrittled and the steel sheet may fracture during rolling. Therefore, the Sn content is set to 1.000% or less. The Sn content is preferably 0.500% or less.

Sb: 0% to 0.200%

Sb is an element that suppresses the coarsening of the structure and contributes to the improvement in the strength. When the Sb content is less than 0.001%, there are cases where the above effect is not sufficiently obtained. Therefore, the Sb content is preferably 0.001% or more. The Sb content is more preferably 0.005% or more.

On the other hand, when the Sb content exceeds 0.200%, the steel sheet may be excessively embrittled and the steel sheet may fracture during rolling. Therefore, the Sb content is set to 0.200% or less. The Sb content is preferably 0.100% or less.

As described above, the steel sheet according to the present embodiment contains essential elements and the remainder consisting of Fe and impurities, or contains essential elements, one or more optional elements, and the remainder consisting of Fe and impurities. Impurities are elements that are unavoidably incorporated from steel raw materials and/or in a steelmaking process, and are elements that are allowed within the range that does not impair the properties of the steel sheet according to the present embodiment.

For example, Ti, Mo, Nb, Cr, B, V, Ni, Cu, W, Ca, Ce, Mg, Zr, La, REM, Sn, and Sb may be treated as impurities when the amounts thereof are all trace amounts lower than the above-mentioned preferable lower limits.

Furthermore, as impurities, in addition to the above elements, H, Na, Cl, Sc, Co, Zn, Ga, Ge, As, Se, Y, Tc, Ru, Rh, Pd, Ag, Cd, In, Te, Cs, Ta, Re, Os, Ir, Pt, Au, and Pb are allowed in a range of 0.010% or less in total.

For the chemical composition of the entire steel sheet, a sample of 1000 mm^3 or more collected from a ¼ thickness to ⅜ thickness position in a sheet width center portion of any of molten steel immediately before casting, a slab, a steel sheet before cold rolling, a steel sheet after cold rolling, or a steel sheet after an annealing step is sampled. This sample is obtained by analysis by inductively coupled plasma (ICP) atomic emission spectrometry.

Next, the microstructure of the steel sheet according to the present embodiment will be described.

The steel sheet according to the present embodiment contains ferrite, martensite and residual austenite, bainite, and pearlite, and the area fractions there are limited. By forming such a structure, a steel sheet having high strength, high work hardening, and excellent formability, particularly ductility, is obtained.

Furthermore, in the steel sheet according to the present embodiment, macrosegregation of Mn is relaxed to reduce a change in the amount of work hardening depending on the location of the steel sheet, and microsegregation of Mn is strengthened, whereby a steel sheet having low yield strength and excellent formability is obtained.

Next, the microstructure of the steel sheet according to the present embodiment will be described.

In the steel sheet according to the present embodiment, the microstructure in a range between a ⅛ thickness position (⅛ thickness) in a sheet thickness direction from the surface of the steel sheet and a ⅜ thickness position (⅜ thickness) in the sheet thickness direction from the surface with a ¼ thickness (¼ thickness) position in the sheet thickness direction from the surface of the steel sheet as the center is limited. The reason for this is that the microstructure between the ⅛ thickness and the ⅜ thickness with the ¼ thickness in the sheet thickness direction from the surface of the steel sheet as a center position in the sheet thickness direction is a representative structure of the steel sheet, and the configuration thereof correlates with the mechanical properties of the entire steel sheet. Therefore, in the present embodiment, the range in the sheet thickness direction for specifying the microstructural fraction is set to "the ⅛ thickness to the ⅜ thickness with the ¼ thickness as the center position in the sheet thickness direction". In addition, "%" in a case of expressing the microstructural fraction is an area fraction.

<Ferrite: 10% to 97%>

Ferrite is a structure that contributes to the improvement in the ductility. When the ferrite fraction is less than 10%, the ductility decreases. Therefore, the ferrite fraction is set to 10% or more. The ferrite fraction is preferably 15% or more.

On the other hand, since it is difficult to increase the strength to 590 MPa or more with ferrite alone, it is necessary to contain 3% or more of residual austenite and martensite, which will be described later. Therefore, the ferrite fraction is set to 97% or less. The ferrite fraction is preferably 95% or less.

The ferrite mentioned here includes both recrystallized ferrite and unrecrystallized ferrite.

<Residual Austenite and Martensite: 3% to 90%>

In order to secure the strength, the steel sheet according to the present embodiment needs to contain residual austenite and martensite in a total amount of 3% to 90%. Residual austenite transforms into martensite when worked, thus increases the strength like martensite. In addition, residual austenite can also increase the ductility by transformation-induced plasticity. The total area fraction (total fraction) of residual austenite and martensite is preferably 5% or more.

On the other hand, if the fraction thereof is too high, the ductility decreases. Therefore, the total area fraction of residual austenite and martensite is preferably set to 90% or less. The total area fraction of residual austenite and martensite is more preferably 85% or less.

<Bainite: 0% to 87%>

Bainite has higher strength than ferrite, and thus can contribute to high-strengthening. However, since bainite has low ductility, the upper limit of the bainite fraction is set to 87% in consideration of the balance between strength and formability. The bainite fraction is preferably 85% or less. Since the strength can be improved by residual austenite and martensite, bainite may not be contained, and the lower limit of the area fraction (fraction) of bainite is 0%. However, the area fraction of bainite may be, for example, 5% or more.

<Pearlite: 0% to 10%>

Pearlite is a composite structure of cementite and ferrite. Pearlite not only significantly deteriorates toughness, but also does not increase work hardening, so that the effect thereof on an increase in ductility and strength is small. Therefore, the pearlite fraction is limited to 10% or less. The pearlite fraction is preferably 5% or less.

The pearlite may not be contained in the steel sheet according to the present embodiment, and the lower limit of the pearlite fraction is 0%. However, the pearlite fraction may be, for example, 2% or more.

The area fraction of each phase is calculated by the following method.

A sample with a sheet thickness direction cross section parallel to a rolling direction of the steel sheet as an observed section is collected, and the observed section is polished and subjected to nital etching. The observed section after the nital etching is observed with an optical microscope or a scanning electron microscope (SEM). The area fraction of each structure is calculated by a taken image or an image analysis software in the device. One visual field in the image is set to 200 μm in length and 200 μm or more in width, the area fraction of each structure is calculated by performing image analysis for each of 10 or more different visual fields, the average value thereof is obtained, and the average value is determined to be the area fraction.

A flat region that is recessed from the surrounding structure such as martensite, has no lower structure, and has few irregularities is determined to be ferrite.

Since pearlite presents a lamellar structure in which ferrite and cementite are layered, the lamellar region is determined to be pearlite. Pseudo-pearlite with layered cementite that is cut in the middle is also included in pearlite.

Bainite is recessed from the martensite structure like ferrite, but is different from ferrite. Bainite has a morphology with elongated laths or a block-shaped morphology, and has carbides and residual austenite present between laths and blocks. Therefore, a structure having a morphology with elongated laths or a block-shaped morphology, and has carbides and residual austenite present between laths and blocks is determined to be bainite.

In the steel sheet according to the present embodiment, after identifying ferrite, pearlite, and bainite based on the above features and measuring the area fractions thereof, regions other than ferrite, bainite, and pearlite are determined to be martensite and residual austenite, and the area fractions thereof are obtained.

Since both martensite and residual austenite have flat structures when observed with the SEM, distinguishment therebetween is difficult. However, since austenite transforms into martensite after being worked, it is not necessary to specify the area fraction of each of residual austenite and martensite, and the total area fraction thereof is specified.

The structure is measured at the ¼ thickness position (¼ thickness) from the surface of the sheet thickness direction cross section parallel to the rolling direction as the center.

<Average Distance Between Centers of Adjacent High Mn Regions in Plane Parallel to Rolled Surface at ¼ Thickness is 1.00 mm or Shorter>

The present inventors found that in a plane which is parallel to the rolling direction of the steel sheet and a plane which is perpendicular to the sheet thickness direction (a plane parallel to the surface of the steel sheet), a region having a high Mn content and a region having a low Mn content are different in work hardening ability, and this difference leads to a difference in work hardening ability depending on the location of the steel sheet.

In addition, the present inventors further examined in a microscopic view in what degree of region the Mn content has to be uniform to reduce the difference in work hardening ability depending on the location. As a result, for example, in a finite element method (FEM) analysis used in a case of predicting a deformation behavior during press forming, when each element was set to a region of about several mm×several mm, the prediction accuracy of forming increased. From this, it was found that the steel sheet may be uniform on the order of several mm.

Therefore, the present inventors set a measurement range to a region of 100 mm×100 mm, and when the maximum value of Mn contents in the measurement range was indicated as Mnmax and the average value of the Mn contents in the measurement range was indicated as Mnave, determined regions having a Mn content of (Mnave+Mnmax)/2 or more to be high Mn regions and regions other than the high Mn regions to be low Mn regions, and examined an effect of the presence state of the high Mn regions on the work hardening ability.

As a result, it was found that the difference in the amount of work hardening is decreased when the structure is uniform to a degree such that the average of distances between the centers of the high Mn regions (regions having a high Mn content) adjacent to each other (the distance between the center of the high Mn region and the center of the high Mn region adjacent thereto) is 1.00 mm or shorter.

Therefore, in the steel sheet according to the present embodiment, the average distance between the centers of the high Mn regions in the plane parallel to the surface of the steel sheet at the ¼ thickness is set to 1.00 mm or shorter. The average distance between the centers of the high Mn regions in the plane parallel to the surface of the steel sheet at the ¼ thickness is preferably 0.80 mm or shorter.

On the other hand, when the average distance between the centers of the high Mn regions in the plane parallel to the surface of the steel sheet at the ¼ thickness is shorter than 0.15 mm, the sizes of individual high Mn regions become fine and the deviation of the number densities of the high Mn regions at each location increases, so that there is concern that the accuracy of press forming may deteriorate. Therefore, the average distance between the centers of the high Mn regions is preferably set to 0.15 mm or longer. The average distance between the centers of the high Mn regions in the plane parallel to the surface of the steel sheet at the ¼ thickness is more preferably 0.25 mm or longer.

The high Mn regions and the low Mn regions can be determined by an electron probe micro analyzer (EPMA).

Specifically, the plane parallel to the surface of the steel sheet is polished from the surface of the steel sheet to the ¼ thickness position (¼ thickness), and the distribution of Mn in a region of 100 mm×100 mm is obtained by the EPMA. Since the absolute value of the Mn concentration (Mn content) is important, a standard sample of C, Mn, and other contained elements is measured and quantitatively analyzed. The interval between points to be measured may be appropriately adjusted depending on a measurement time and the like. However, since the region having a high Mn concentration has a size of about 0.2 to 0.3 mm, the measurement interval is preferably 0.01 mm or shorter. In the examination in the present embodiment, the measurement interval is set to 0.01 mm.

After the measurement, the average value of the Mn contents of the entire measurement region is indicated as Mnave, the maximum value thereof is indicated as Mnmax, and a region having a Mn content of (Mnave+Mnmax)/2 or more is determined to be the high Mn region. More specifically, in a case where ten or more measurement points having a Mn content of (Mnave+Mnmax)/2 or more are continuously connected and a plane can be specified by these points, the region surrounded by these measurement points is determined to be the high Mn region. There are cases where several measurement points lower than (Mnave+Mnmax)/2 are included inside the surrounded region, but the measurement points are also a portion of the high Mn region.

Then, the centers of all the measured high Mn regions are obtained. The distance from the center of each high Mn region to the center of the adjacent high Mn region is obtained, and the average thereof is determined to be the average distance between the centers of the high Mn regions. The center is obtained from the coordinates of each of the high Mn regions measured by the EPMA. For example, when n measurement points are included in the high Mn region, each measurement point is numbered from 1 to n, and each coordinate is defined as (X_i,Y_i) (i is an integer from 1 to n), the center coordinates (Xc,Yc) are defined as $(\{X_1+X_2+ \ldots +X_n\}/n, \{Y_1+Y_2+ \ldots +Y_n\}/n)$. $X_1+X_2+\ldots+X_n$ means that all n coordinates from X_1 to X_n are added.

<Density (Number Density) D_A of High Mn Regions at Sheet Width Center Portion and Density (Number Density) D_B of High Mn Regions at ¼ Width Position from Sheet Width End Portion in Plane Parallel to Rolling Direction at ¼ Thickness Position in Sheet Thickness Direction from Surface Satisfy $0.77 \le D_A/D_B \le 1.30$>

As described above, even if the average distance between the centers of the high Mn regions is short and uniform in the region in units of several mm, in a case where the densities of the high Mn regions vary in regions separated by several hundred mm, the accuracy of press forming decreases.

As a result of the examination by the present inventors, it was found that in order to reduce the difference in the amount of work hardening between a center portion in a sheet width direction (a w/2 position from a sheet width end portion when the sheet width is indicated as w) and a ¼ width position (¼ width) from the center portion, when the difference between a density D_A of the high Mn regions at the sheet width center portion and a density D_B of the high Mn regions at the ¼ width is reduced, the difference in the amount of work hardening in the width direction can be reduced. Specifically, it was found that when the density (number density) D_A of the high Mn regions at the sheet width center portion and the density (number density) D_B of the high Mn regions at the ¼ width satisfy Expression (1), the difference in the amount of work hardening in the width direction can be sufficiently reduced. When D_A/D_B is outside the range of Expression (1), the difference in the amount of work hardening in the width direction increases. D_A/D_B preferably satisfies the range of Expression (2), and more preferably satisfies the range of Expression (3).

$$0.77 \le D_A/D_B \le 1.30 \quad \text{Expression (1)}$$

$$0.80 \le D_A/D_B \le 1.25 \quad \text{Expression (2)}$$

$$0.83 \le D_A/D_B \le 1.20 \quad \text{Expression (3)}$$

The density (density D_A and density D_B) of the high Mn regions at each position of the sheet width center portion and the ¼ width is obtained by measuring the measurement region of 100 mm×100 mm with the EPMA. Here, the measurement region may be determined so that the center line of the measurement region parallel to the rolling direction substantially coincides with the sheet width center portion and the distance of ¼ from one end in the sheet width direction. The density of the high Mn regions is the number density of the high Mn regions per area of entire each measurement region (100 mm×100 mm), and is expressed in units of, for example, $/mm^2$.

<Ratio of Average Hardness of High Mn Regions to Average Hardness of Low Mn Regions in Plane Parallel to Rolling Direction at ¼ Thickness Position in Sheet Thickness Direction from Surface is 1.1 to 2.0>

Even if the average distance between the centers of the high Mn regions is 1.00 mm or shorter and the ratio between the densities of the high Mn regions at the width center portion and the ¼ width is within the range satisfying Expression (1), in a case where the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions exceeds 2.0, the change in work hardening depending on the location of the steel sheet increases. Therefore, the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions is set to 2.0 or less. The ratio is preferably 1.90 or less, and more preferably 1.80 or less.

On the other hand, although it is not necessary to particularly determine the lower limit of the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions, since there is a difference in Mn content between the high Mn regions and the low Mn regions, the ratio usually becomes 1.1 or more. Therefore, the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions in the plane parallel to the rolled surface at the ¼ thickness position in the sheet thickness direction from the surface is set to 1.1 or more.

The hardness is measured according to the Vickers hardness test JISZ 2244:2009. The load is set to a degree at which the indentation becomes several μm, for example, 10 gf, and a region of 400 μm×400 μm is measured at a pitch of 0.2 μm. Then, the average hardness of the high Mn regions and the average hardness of the low Mn regions are calculated.

<Difference Between Average of Top 5% and Average of Bottom 5% of Mn Contents at Measurement Points in Low Mn Regions in Plane Parallel to Rolling Direction at ¼ Thickness Position in Sheet Thickness Direction from Surface is 0.3 Mass % or More>

The macrosegregation of Mn (on the order of 100 μm to several mm) can be reduced by controlling the average distance between the centers of the high Mn regions, the ratio between the density D_A of the high Mn regions at the sheet width center portion and the density D_B of the high Mn regions at the ¼ width position from the sheet width end portion, and the ratio between the average hardness of the high Mn regions and the average hardness of the low Mn regions. As a result, for example, the change in the amount of work hardening depending on the location in the width direction can be reduced.

On the other hand, in a case where the above control is simply performed, microsegregation portions also decrease in size. However, it is preferable that the microsegregation is at a certain level or higher. This is because the presence of the microsegregation portions allows Mn to be concentrated, and as a result, the change in the amount of work hardening depending on the location in the rolling direction, which is perpendicular to the width direction, can be reduced. Although the reason for this is not clear, for example, it is considered that as Mn is concentrated, a temperature range of Ac1 to Ac3 during a heat treatment is widened, so that a slight change in microstructure with respect to a slight temperature change during the heat treatment further decreases and even a slight change in mechanical properties, which is only a change in the amount of work hardening, decreases.

As a result of the examination by the present inventors, it was found that when the microsegregation portions are small, the change in the amount of work hardening with respect to the position in the rolling direction increases, and this tendency is significant in a case where the difference in the average value of the top 5% and the average value of the bottom 5% of the Mn contents of the low Mn regions is less than 0.3 mass %.

Therefore, in the steel sheet according to the present embodiment, the difference between the average of the top 5% and the average of the bottom 5% of the Mn contents at measurement points in the measurement range of the low Mn regions is set to 0.3 mass % or more. The difference between the average of the top 5% and the average of the bottom 5% of the Mn contents at the measurement points in the measurement range of the low Mn regions is preferably 0.4 mass % or more.

On the other hand, in a case where the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents at the measurement points in the low Mn regions in the measurement range is excessively large, a brittle region serving as an origin of fracture is formed, and there is concern that the formability of the steel sheet may decrease. Therefore, the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents at the measurement points in the low Mn regions in the measurement range is preferably 1.00 mass % or less.

The difference between the average of the top 5% and the average of the bottom 5% of the Mn contents at the measurement points in the low Mn regions is determined by the following method. Among the low Mn regions obtained by the EPMA measurement, in a region 20 μm or longer away from the end portion of the high Mn region, a region of 200 μm×200 μm is measured with the EPMA at a step (interval) of 0.05 μm. For a steel that cannot secure the low Mn region of 200 μm×200 μm at the position 20 μm or longer away from the end portion of the high Mn region, the Mn contents of a plurality of low Mn regions at the position 20 μm away from the end portion of the high Mn region are measured so that the total area becomes 40,000 μm^2 or more. The average value of the top 5% and the average value of the bottom 5% of the Mn contents of the obtained measurement points are obtained, and the difference is obtained.

The surface of the steel sheet according to the present embodiment may be hot-dip galvanized. That is, the steel sheet according to the present embodiment may a hot-dip galvanized steel sheet having a hot-dip galvanized layer on its surface. Corrosion resistance can be improved by hot-dip galvanizing. The hot-dip galvanized layer may be a hot-dip galvannealed layer. The hot-dip galvanized layer may be a hot-dip galvanized layer formed under normal plating conditions (including a plating layer formed by hot-dip plating with a zinc alloy), and the hot-dip galvannealed layer may be a plating layer obtained by alloying a hot-dip galvanized layer under normal alloying treatment conditions.

When the galvanized layer is a hot-dip galvannealed layer, in addition to the corrosion resistance, the number of continuous spots that can be formed during spot welding increases compared to a case where alloying is not performed.

The plating adhesion amount of the hot-dip galvannealed layer is not limited to a specific amount, but is preferably 5 g/m^2 or more per surface in terms of securing the required corrosion resistance.

In the steel sheet of the present embodiment, upper layer plating (for example, Ni plating) may be applied onto the hot-dip galvannealed layer for the purpose of improving coatability and weldability. Furthermore, various treatments such as a chromate treatment, a phosphate treatment, a lubricity improvement treatment, and a weldability improvement treatment may be performed for the purpose of improving the surface properties of the hot-dip galvannealed layer.

The sheet thickness of the steel sheet according to the present embodiment is not limited, but is preferably 0.1 to 11.0 mm. A steel sheet having a sheet thickness of 0.1 to 11.0 mm is suitable as a base steel sheet for a member for a vehicle produced by press working. In addition, a high strength thin steel sheet having the above-mentioned sheet thickness can be easily produced on a thin sheet production line.

Next, a method for producing the steel sheet according to the present embodiment will be described.

The steel sheet according to the present embodiment achieves its effects regardless of the production method as long as the steel sheet has the above features, and can be stably produced by a production method including the following steps, which is preferable.

(I) A casting step of producing a slab by melting a steel having the same chemical composition as the steel sheet according to the present embodiment described above, casting the melted steel to produce a slab, and cooling the slab at a temperature of 950° C. to 550° C. while applying a pressure of 10 N/cm^2 or more to the slab in a thickness direction so that an average cooling rate is 100° C./h or faster.

(II) A heating step of heating the slab to a temperature range of 1100° C. to 1280° C. after cooling the slab to room temperature or before cooling the slab to room temperature so that a heating rate in a temperature range of 650° C. to 850° C. is 50° C./min or slower.

(III) A hot rolling step of hot-rolling the slab after the heating step in a temperature range of 1050° C. or higher at a cumulative rolling reduction of 35% or more to obtain a hot-rolled steel sheet.

(IV) A cooling step of cooling the hot-rolled steel sheet to 650° C. or lower, the cooling being started within three seconds after the hot rolling step is completed, so that an average cooling rate to 700° C. is 20° C./s or faster.

(V) A coiling step of coiling the hot-rolled steel sheet after the cooling step in a temperature range of 300° C. to 650° C.

(VI) A pickling step of performing pickling on the hot-rolled steel sheet after the coiling step to obtain a pickled steel sheet.

(VII) A cold rolling step of performing cold rolling on the pickled steel sheet to obtain a cold-rolled steel sheet.

(VIII) An annealing step of heating the cold-rolled steel sheet to an annealing temperature of Ac1° C. to 1000° C. at an average heating rate of 10.0° C./s or slower and performing holding at the annealing temperature for five seconds to 600 seconds.

(IX) A post-annealing cooling step of cooling the cold-rolled steel sheet after the annealing step to a retention temperature of 150° C. to 550° C. at an average cooling rate of 1° C./s to 200° C./s.

(X) A retaining step of performing retention at the retention temperature for 15 seconds to 1000 seconds.

(XI) A final cooling step of cooling the cold-rolled steel sheet after the retaining step to room temperature.

Hereinafter, each step will be described.

<Casting Step>

In the casting step, a slab is produced by melting a steel having the same composition as the steel sheet according to the present embodiment, casting the melted steel to produce a slab, and cooling the slab at a temperature of 950° C. to 550° C. while applying a pressure of 10 N/cm^2 or more to the slab in a thickness direction so that an average cooling rate is 100° C./h or faster. As for a cooling method in a cooling process until solidification, it is preferable to perform cooling while performing adjustment such as increasing the amount of water in the center portion so that the cooling rate becomes constant in the width direction. The thickness direction mentioned here is a direction corresponding to the sheet thickness direction of the steel sheet after the hot rolling step which is a subsequent step.

At 950° C. to 550° C., Mn concentration occurs. When the cooling rate in this temperature range is slow, the average distance between the centers of the high Mn regions in the plane at the ¼ thickness (the plane parallel to the rolling direction at the ¼ thickness position in the sheet thickness direction from the surface of the steel sheet) may exceed 1.00 mm, Expression (1) may not be satisfied, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Although the cause of this is not clear, at 950° C. to 550° C. in the cooling process, phases constituting the structure are divided into a plurality of phases such as ferrite and austenite. This structure becomes a non-uniform structure reflecting unevenness of the concentration of Mn and the like generated during casting, and Mn is likely to be concentrated. Therefore, it is considered that this is because the Mn content in the high Mn region is increased and the high Mn regions are easily localized. When the average cooling rate is slower than 100° C./h, deviation from the ranges of the present invention is incurred as described above. Therefore, the average cooling rate at 950° C. to 550° C. is set to 100° C./h or faster. The average cooling rate at 950° C. to 550° C. is preferably 150° C./h or faster.

On the other hand, when the average cooling rate is excessively fast, a temperature deviation inside the cast piece increases, and there is a risk that the slab will crack. Therefore, the average cooling rate at 950° C. to 550° C. is preferably 500° C./h or slower.

In the casting step, it is necessary to cool the slab at a temperature of 950° C. to 550° C. while applying a pressure of 10 N/cm^2 or more in the thickness direction to the slab. In a case where no pressure is applied or the pressure is low, the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, Expression (1) may not be satisfied, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Although this mechanism is not clear, it is considered that the crystal lattice is compressed and the diffusion of C and Mn is slowed down by the application of the pressure. Therefore, the pressure applied to the slab in the thickness direction is set to 10 N/cm^2 or more. The pressure applied to the slab in the thickness direction at a temperature of 950° C. to 550° C. is preferably 30 N/cm^2 or more.

On the other hand, when an excessively high pressure is applied, the slab is deformed and there is a risk that the slab will crack. Therefore, in the temperature range of 950° C. to 550° C., the pressure applied to the slab in the thickness direction is preferably set to 100 N/cm^2 or less.

<Heating Step>

The cast slab is heated to a temperature of 1100° C. to 1280° C. after cooling the slab to room temperature or before cooling the slab to room temperature. When the heating temperature is too low, carbides may remain undissolved. In this case, even if a subsequent heat treatment is performed, C is contained in the carbides, so that the fraction of martensite and residual austenite that require a large amount of C is reduced. Therefore, the lower limit is set to 1100° C. The heating temperature is preferably 1190° C. or higher. On the other hand, when the heating temperature is too high, the production cost increases, the grain size excessively increases, and the toughness decreases. Therefore, the upper limit is set to 1280° C. The heating temperature is preferably 1275° C. or lower.

In addition, in the heating step, the average heating rate in a temperature range of 650° C. to 850° C. is set to 50° C./min or slower.

By such heating, microsegregation proceeds, and the difference in the amount of work hardening in the rolling direction can be reduced. When the average heating rate exceeds 50° C./min, the microsegregation insufficiently proceeds, and the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents in the low Mn regions decreases. The average heating rate in the temperature range of 650° C. to 850° C. is preferably 40° C./min or slower.

On the other hand, retention in the heating step for an excessively long time impairs the surface quality of the slab and deteriorates the external appearance of a final product, which is not preferable. From this viewpoint, the average heating rate in the temperature range of 650° C. to 850° C. is preferably set to 5° C./min or faster. The average heating rate in the temperature range of 650° C. to 850° C. is more preferably 10° C./min or faster.

<Hot Rolling Step>

The heated slab is hot-rolled in a temperature range of 1050° C. or higher at a cumulative rolling reduction of 35% or more to obtain a hot-rolled steel sheet. Macrosegregation can be reduced by large deformation due to hot rolling. When the cumulative rolling reduction in the temperature range of 1050° C. or higher is less than 35%, the effect of suppressing macrosegregation decreases, so that the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, Expression (1) may not be satisfied, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Therefore, the lower limit of the cumulative rolling reduction at 1050° C. or higher is set to 35%. The cumulative rolling reduction in the temperature range of 1050° C. or higher is preferably 40% or more. On the other hand, the cumulative rolling reduction at 1050° C. or higher is not particularly limited and may be, for example, 80% or less.

Even if the rolling reduction at lower than 1050° C. is increased, the effect of reducing macrosegregation is small. Therefore, the cumulative rolling reduction in the temperature range of 1050° C. or higher is controlled.

<Cooling Step>

Cooling (rapid cooling) is started within three seconds after the hot rolling is ended. By shortening the time until the start of cooling, macrosegregation can be reduced. When the time from the end of the hot rolling to the start of the cooling exceeds three seconds, the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, Expression (1) may not be satisfied, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Although this mechanism is not clear, it is presumed that the Mn dispersed by the hot rolling is concentrated again when subsequently held at a high temperature. The hot rolling end time point is ended refers to the time point at which rolling by the final rolling roll in the hot rolling step is ended. In addition, the cooling start time point refers to the time point at which cooling is started at a cooling rate of 10° C./sec or faster by spraying a cooling medium such as water. As the cooling medium, water, gas water, a gas such as nitrogen gas, hydrogen gas, helium gas, or air, or a mixture thereof may be used.

In the cooling step, cooling to 650° C. or lower is performed by setting the average cooling rate from the start of the cooling to 700° C. to 20° C./s or faster. When the average cooling rate is slower than 20° C./s, the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, Expression (1) may not be satisfied, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Although this mechanism is not clear, it is presumed that this is because coarse ferrite tends to be generated in a non-uniform manner in the temperature range. When the average cooling rate is slow, transformation into coarse ferrite proceeds and the microstructure becomes non-uniform. In a non-uniform structure, the amount of ferrite tends to vary greatly depending on the position in the sheet width direction. It is considered that since the concentration of Mn from ferrite to austenite occurs, the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. It is considered that when the amount of ferrite is different, the degree of concentration is also different, so that the degree of the concentration of Mn in the width direction is different, and Expression (1) is not satisfied.

The lower limit of the cooling temperature in the cooling step is preferably 300° C. When the cooling temperature is 300° C. or higher, it is not necessary to reheat the hot-rolled steel sheet after the cooling step before the coiling step which is the next step.

<Coiling Step>

In the coiling step, the hot-rolled steel sheet after being cooled to a temperature range of 650° C. or lower is coiled in a temperature range of 300° C. to 650° C. As described above, the coiling treatment is performed after forming a uniform structure by controlling the average cooling rate to 700° C. in the cooling step. When the coiling temperature exceeds 650° C., ferrite grows coarsely, so that the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, Expression (1) may not be satisfied, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Although this mechanism is not clear, it is considered that coarse ferrite is likely to be generated even in the temperature range, a non-uniform structure is likely to be formed, and thus Mn is non-uniformly concentrated, so that the average distance between the centers of the high Mn regions in the plane at the ¼ thickness may exceed 1.00 mm, or the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions may be outside the above range. Therefore, the coiling temperature is set to 650° C. or lower. The coiling temperature is more preferably 600° C. or lower.

On the other hand, when the coiling temperature is lower than 300° C., microsegregation does not proceed, and even in a region near the sheet width center portion, the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn concentrations in the low Mn regions decreases. Therefore, the coiling temperature is set to 300° C. or higher. The coiling temperature is preferably 460° C. or higher.

<Pickling Step>

In the pickling step, the hot-rolled steel sheet after the coiling step is coiled again as necessary, and then subjected to pickling to obtain a pickled steel sheet. The pickling conditions may be set according to a normal method. In a case where a large amount of scale is adhered, the concentration of hydrochloric acid or the like may be increased or the temperature may be increased.

<Cold Rolling Step>

In the cold rolling step, the pickled steel sheet is cold-rolled to obtain a cold-rolled steel sheet having a predetermined sheet thickness. Cold rolling conditions such as a rolling reduction may be a normal method.

<Leveling Step>

After the cold rolling step, it is preferable to perform working on the cold-rolled steel sheet using a leveler. Leveling is not essential, but leveling causes a decrease in the absolute value of D_A/D_B in Expression (1), which is preferable. Although the reason for this is not clear, it is considered that when leveling is performed, the sheet shape is improved, strain of the leveler remains uniform in the width direction, and the distribution of Mn in a subsequent heat treatment proceeds or heat is uniformly applied, so that the subsequent diffusion of Mn or the like also uniformly occurs. The leveling conditions are not limited, but a condition under which the amount of strain introduced onto the surface of the steel sheet by a roll leveler becomes 0.2% to 0.3% is preferable.

<Annealing Step>

In the annealing step, the cold-rolled steel sheet is heated to an annealing temperature of Ac1° C. to 1000° C. at an average heating rate of 10.0° C./s or slower, and held at the annealing temperature for five seconds to 600 seconds.

When the average heating rate exceeds 10.0° C./s, microsegregation does not proceed and the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents in the low Mn regions decreases. Although this mechanism is not clear, during heating from room temperature, in the structure, carbides such as cementite are dispersed in the structure having a BCC (body-centered cubic) structure such as ferrite, bainite, and martensite. It is considered that Mn is diffused and concentrated from the BCC structure to cementite dispersed therein and thus the difference in Mn content increases even in the low Mn regions. Therefore, the average heating rate is set to 10.0° C./s or slower. The average heating rate is preferably 8.0° C./s or slower.

When the heating temperature (annealing temperature) is lower than Ac1° C., cementite remains undissolved, so that residual austenite and martensite are reduced, and strength and ductility decrease. Therefore, the heating temperature is set to Ac 1° C. or higher. The heating temperature is preferably 745° C. or higher. Ac1 is obtained by the following expression described in The Physical Metallurgy of Steels, William C. Leslie, p. 273, using the chemical composition.

$$Ac1=723-10.7\times Mn-16.9\times Ni+29.1\times Si+16.9\times Cr+290\times As+6.38\times W$$

Here, in the above expression, Mn, Ni, Si, Cr, As, and W are the amounts (mass %) of the corresponding elements in the steel sheet.

On the other hand, when the heating temperature of the annealing exceeds 1000° C., the ferrite fraction is significantly reduced, and the balance between strength and ductility is deteriorated. Therefore, the heating temperature is set to 1000° C. or lower. The heating temperature is preferably 950° C. or lower.

In a case of holding in the temperature range of Ac 1° C. to 1000° C. for five seconds to 600 seconds, ferrite and cementite are transformed into austenite. When the retention time is shorter than five seconds, the dissolution of cementite does not occur stably, and residual austenite and martensite cannot be obtained. Therefore, the retention time is set to five seconds or longer. The retention time is, preferably 50 seconds or longer. On the other hand, when the retention time is too long, the cost is significantly increased. Therefore, the upper limit thereof is set to 600 seconds. The retention time is preferably 240 seconds or shorter.

<Post-Annealing Cooling Step>

After the annealing step, cooling to a temperature range (retention temperature) of 150° C. to 550° C. at an average cooling rate of 1° C./s to 200° C./s is performed.

When the average cooling rate is slower than 1° C./s, pearlitic transformation proceeds and the strength and ductility decrease. Therefore, the average cooling rate is 1° C./s or faster. The average cooling rate is preferably 5° C./s or faster. On the other hand, when the cooling rate is too fast, an uneven cooling rate occurs in the longitudinal direction and the width direction of the steel sheet, and an uneven volume change occurs accordingly, so that the shape of the sheet deteriorates and stable press forming cannot be performed. Therefore, the average cooling rate is set to 200° C./s or slower. The average cooling rate is preferably 60° C./s or slower.

The reason why the cooling stop temperature is set in the above range is that bainitic transformation occurs at 150° C. to 550° C. and contributes to an increase in the strength.

<Retaining Step>

After the post-annealing cooling step, retention in the temperature range (retention temperature: 550° C. to 150° C.) is performed. Retention means to control the temperature of the cold-rolled steel sheet to be held in a temperature range of 150° C. to 550° C. for a time of 15 seconds to 1000 seconds by holding or cooling the temperature of the cold-rolled steel sheet.

When the retention time is short, the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions excessively increases. By performing holding the martensite in the temperature range of 150° C. to 550° C. for 15 seconds to 1000 seconds, the strength of martensite decreases. It is considered that in the high Mn regions where the amount of martensite is large, the decrease in hardness due to the holding is large, so that the above ratio decreases due to the holding, and as a result, the difference in work hardening depending on the location can be reduced. Therefore, the retention time is set to 15 seconds or longer. The retention time is preferably 35 seconds or longer.

On the other hand, when the retention time is too long, the number of moving dislocations around martensite decreases and the yield ratio increases. When the retention time exceeds 1000 seconds, the yield ratio becomes particularly large. Therefore, the upper limit thereof is set to 1000 seconds. The retention time is more preferably 600 seconds or shorter.

<Final Cooling Step>

The cold-rolled steel sheet after the retaining step is cooled to room temperature. The cooling conditions are not particularly limited. By subsequent skin pass rolling or the like, light shape correction or the like may be performed.

The steel sheet according to the present embodiment can be obtained by the production method including the above steps.

However, in a case where the steel sheet is a hot-dip galvanized steel sheet or a galvannealed steel sheet, it is preferable to further perform the following steps.

<Hot-Dip Galvanizing Step>

In a case where a hot-dip galvanized layer is formed on the surface of the steel sheet, it is preferable that the cold-rolled steel sheet is immersed in a molten zinc bath between the retaining step and the final cooling step. The plating conditions may be set according to a normal method.

<Alloying Step>

After the hot-dip galvanizing step, the cold-rolled steel sheet may be reheated to 470° C. to 550° C. and held for 60 seconds or shorter to form the hot-dip galvanized layer into a hot-dip galvannealed layer, thereby obtaining a galvannealed steel sheet.

EXAMPLES

Next, examples of the present invention will be described. However, the conditions in the examples are one example of conditions adopted to confirm the feasibility and effects of the present invention, and the present invention is not limited to this one example of conditions. The present invention can adopt various conditions as long as the object of the present invention is achieved without departing from the gist of the present invention.

Molten steels having the compositions shown in Table 1 were continuously cast according to a normal method to obtain cast slabs. In Table 1, the chemical compositions of Kind of steel A to T satisfy the chemical composition of the present invention.

On the other hand, the Nb content in Kind of steel aa, the C content in Kind of steel bb, the Si content in Kind of steel cc, the Mn content in Kinds of steel dd and ee, the P content in Kind of steel ff, the S content in Kind of steel gg, the Al content in Kind of steel hh, and the Ti content in Kind of steel ii did not satisfy the ranges of the chemical composition of the present invention.

TABLE 1

Kind of	Chemical composition (mass %) (remainder consists of Fe and impurities)													
steel	C	Si	Mn	P	S	N	Al	Nb	Ti	Mo	Cr	B	Ni	V
A	0.085	0.60	2.08	0.008	0.0027	0.0028	0.027	—	—	—	—	—	—	—
B	0.191	1.19	1.82	0.009	0.0021	0.0028	0.025	—	—	—	—	—	—	—
C	0.071	0.40	1.80	0.008	0.0010	0.0013	0.010	—	—	—	—	—	—	—
D	0.150	1.60	1.60	0.001	0.0006	0.0019	0.010	0.050	—	—	—	—	—	—
E	0.084	0.60	2.06	0.008	0.0027	0.0028	0.027	0.050	0.040	—	—	—	—	—
F	0.180	1.45	1.80	0.018	0.0005	0.0018	0.035	—	—	0.210	—	—	—	—
G	0.220	0.47	1.80	0.005	0.0010	0.0010	0.800	—	—	—	1.10	—	—	—
H	0.300	1.30	2.00	0.040	0.0008	0.0016	0.010	—	—	—	—	0.0015	—	—
I	0.069	0.40	1.75	0.008	0.0010	0.0013	0.010	—	—	—	—	—	—	—
J	0.081	0.60	2.06	0.008	0.0027	0.0028	0.027	—	—	—	—	—	—	—
K	0.195	1.19	1.82	0.009	0.0021	0.0028	0.025	—	—	—	—	—	0.30	—
L	0.074	0.40	1.80	0.008	0.0010	0.0013	0.010	—	—	—	—	—	—	—
M	0.150	1.60	1.20	0.001	0.0006	0.0019	0.010	—	—	—	—	—		0.100
N	0.080	0.60	2.06	0.008	0.0027	0.0028	0.027	—	—	—	—	—	—	—
O	0.175	1.46	1.90	0.018	0.0005	0.0018	0.035	—	—	—	—	—	—	—
P	0.220	0.47	1.80	0.005	0.0010	0.0010	0.800	—	—	—	—	—	—	—
Q	0.080	0.60	2.06	0.008	0.0027	0.0028	0.027	—	—	—	—	—	—	—
R	0.192	1.25	1.91	0.009	0.0021	0.0028	0.025	—	—	—	—	—	—	—
S	0.071	0.40	1.80	0.008	0.0010	0.0013	0.010	—	—	—	—	—	—	—
T	0.151	1.60	1.60	0.001	0.0006	0.0019	0.010	—	—	—	—	—	—	—
aa	0.071	0.40	1.80	0.008	0.0010	0.0013	0.010	0.300	—	—	—	—	—	—
bb	0.030	0.70	2.80	0.040	0.0005	0.0018	0.010	—	—	—	—	—	—	—
cc	0.240	3.00	3.30	0.040	0.0009	0.0010	0.010	—	—	—	—	—	—	—
dd	0.080	0.06	4.50	0.020	0.0008	0.0011	0.010	—	—	—	—	—	—	—
ee	0.078	0.06	0.05	0.020	0.0009	0.0016	0.010	—	—	—	—	—	—	—
ff	0.085	0.06	2.90	0.120	0.0009	0.0016	0.010	—	—	—	—	—	—	—
gg	0.084	0.06	1.80	0.020	0.0120	0.0010	0.010	—	—	—	—	—	—	—
hh	0.086	0.06	2.00	0.010	0.0005	0.0011	1.800	—	—	—	—	—	—	—
ii	0.083	0.06	1.60	0.020	0.0009	0.0016	0.010	—	0.400	—	—	—	—	—
U	0.081	0.03	1.35	0.013	0.0041	0.0045	0.452	—	—	0.093	0.62	—	—	—
V	0.123	0.09	2.42	0.009	0.0034	0.0066	1.184	—	0.019	—	—	0.0032	—	—
W	0.057	0.13	1.64	0.012	0.0013	0.0053	0.215	0.013	—	0.251	—	—	—	—
X	0.134	0.32	0.93	0.008	0.0042	0.0041	0.097	—	0.028	0.056	0.41	0.0012	0.35	—
Y	0.102	0.73	3.17	0.016	0.0028	0.0047	0.028	0.023	0.013	—	—	—	—	—
Z	0.077	0.35	2.82	0.012	0.0015	0.0027	0.073	—	0.032	—	0.16	—	—	—

Kind of	Chemical composition (mass %) (remainder consists of Fe and impurities)				
steel	W	Cu	Others	Ac1	Note
A	—	—	—	718	Invention Steel
B	—	—	—	738	Invention Steel
C	—	—	—	715	Invention Steel
D	—	—	—	752	Invention Steel
E	—	—	—	718	Invention Steel
F	—	—	—	746	Invention Steel
G	—	—	—	736	Invention Steel
H	—	—	—	739	Invention Steel
I	—	—	—	716	Invention Steel
J	—	—	Mg: 0.002	718	Invention Steel
K	—	0.50	—	733	Invention Steel
L	—	—	Ca: 0.0020	715	Invention Steel
M	—	—	—	757	Invention Steel
N	0.11	—	—	719	Invention Steel
O	—	—	Ce: 0.0025	745	Invention Steel
P	—	—	Zr: 0.0040	717	Invention Steel
Q	—	—	La: 0.0025	718	Invention Steel
R	—	—	REM: 0.0027	739	Invention Steel
S	—	—	Sn: 0.100	715	Invention Steel
T	—	—	Sb: 0.200	752	Invention Steel
aa	—	—	—	715	Comparative Steel
bb	—	—	—	713	Comparative Steel
cc	—	—	—	775	Comparative Steel
dd	—	—	—	677	Comparative Steel
ee	—	—	—	724	Comparative Steel
ff	—	—	—	694	Comparative Steel
gg	—	—	—	705	Comparative Steel

TABLE 1-continued

hh	—	—	—	703	Comparative Steel
ii	—	—	—	708	Comparative Steel
U	—	—	—	720	Invention Steel
V	—	—	Y: 0.0017	700	Invention Steel
W	—	—	Sc: 0.0015	709	Invention Steel
X	—	0.07	—	723	Invention Steel
Y	—	—	—	710	Invention Steel
Z	—	—	—	706	Invention Steel

The cast slabs having the chemical compositions shown in Table 1 were heated, hot-rolled, cooled, coiled, pickled, subjected to cold rolling, and then subjected to a leveling treatment as necessary under the conditions shown in Tables 2-1 to 2-4. The sheet thickness after cold rolling was set to 0.35 to 1.2 mm. The steel sheets were annealed and cooled under the conditions shown in Tables 2-1 to 2-4. Depending on the conditions, hot dip galvanizing was further performed. In addition, some of the steel sheets were subjected to an alloying treatment. In the pickling, the steel sheet cooled to room temperature was immersed in 5 to 10 mass % hydrochloric acid as hydrogen chloride whose temperature was controlled to 80° C. to 90° C. for a total of 30 seconds to 100 seconds, whereby scale on the surface was removed.

In addition, the cooling temperature in the cooling step in the present example is equal to the coiling temperature in the coiling step. That is, in the cooling step, the hot-rolled steel sheet was cooled to the coiling temperature shown in Tables 2-1 to 2-4.

In Tables 2-1 to 2-4, in Kind of plating of Hot-dip galvanizing step, "GI" indicates hot-dip galvanizing, and "GA" indicates hot-dip galvannealing.

TABLE 2-1

		Casting step		Heating step		Hot rolling step	Cooling step			Cold rolling step		Leveling step
Treatment No.	Kind of steel	Average cooling rate at 950° C. to 550° C. in cooling of slab ° C./h	Pressure applied at 950° C. to 550° C. N/cm²	Average heating rate in temperature range of 650° C. to 850° C. ° C./min	Heating temperature ° C.	Cumulative rolling reduction in temperature range of 1050° C. or higher %	Time until start of rapid cooling after completion of hot railing step sec	Average cooling rate from start of cooling to 700° C. ° C./s	Coiling step Coiling temperature ° C.	Cold rolling reduction %	Sheet thickness after cold rolling mm	Presence or absence of leveling
1	A	158	35	35	1240	41	2	96	490	52	1.20	Present
18	C	50	35	35	1240	40	1	92	550	52	1.20	Present
37	D	60	35	35	1240	43	1	94	530	52	1.20	Present
2	A	50	35	35	1240	40	1	90	530	52	1.20	Present
3	A	120	35	35	1240	41	2	95	560	52	1.20	Present
86	E	164	0	35	1240	43	2	98	540	52	1.20	Present
87	F	157	5	35	1240	40	2	98	560	52	1.20	Present
4	A	161	0	35	1240	43	2	95	570	52	1.20	Present
5	A	158	15	35	1240	40	1	95	520	52	1.20	Present
90	G	161	35	70	1240	42	1	99	470	52	1.20	Present
91	H	161	35	60	1240	40	2	99	530	52	1.20	Present
6	A	163	35	60	1240	45	1	90	540	52	1.20	Present
7	A	157	35	45	1240	45	2	100	520	80	0.35	Present
8	A	165	35	35	1050	45	2	90	530	52	1.20	Present
10	A	159	35	35	1240	25	1	100	490	52	1.20	Present
11	A	159	35	35	1240	43	5	99	490	52	1.20	Present
12	A	155	35	35	1240	40	10	98	480	52	1.20	Present
13	A	155	35	35	1240	43	1	10	470	52	1.20	Present
14	A	164	35	35	1240	45	2	15	550	52	1.20	Present
15	A	162	35	35	1240	44	1	30	450	52	1.20	Present
16	A	157	35	35	1240	44	2	95	670	52	1.20	Present
119	A	157	35	35	1240	44	2	95	280	52	1.20	Present
17	A	156	35	35	1240	40	2	100	520	52	1.20	Present
122	C	159	35	35	1240	40	1	92	550	78	0.40	Present
19	A	165	35	35	1240	42	1	95	540	52	1.20	Present
20	A	156	35	35	1240	40	2	95	520	52	1.20	Present

TABLE 2-1-continued

Treatment No.	Annealing step: Average heating rate during heating to Ac1 or higher ° C./s	Annealing step: Annealing temperature ° C.	Annealing step: AC1 ° C.	Annealing step: Retention time sec	Post-annealing cooling step: Average cooling rate from annealing temperature to 550° C. or lower ° C./s	Retaining step: Retention time in temperature range of 150° C. to 550° C. sec	Hot-dip galvanizing step: Presence or absence of plating treatment	Hot-dip galvanizing step: Kind of plating	Alloying step: Reheating (alloying) temperature ° C.	Alloying step: Retention time sec	Note
1	4.4	740	718	110	29	394	Absent	—	—	—	Invention Steel
18	3.0	810	715	110	26	170	Absent	—	—	—	Comparative Steel
37	4.4	820	752	220	55	290	Absent	—	—	—	Comparative Steel
2	3.3	760	718	220	32	229	Absent	—	—	—	Comparative Steel
3	4.4	770	718	110	39	476	Absent	—	—	—	Invention Steel
86	4.2	820	718	160	54	276	Absent	—	—	—	Comparative Steel
87	4.1	780	746	230	53	321	Absent	—	—	—	Comparative Steel
4	3.2	740	718	190	53	378	Absent	—	—	—	Comparative Steel
5	4.7	750	718	100	24	402	Absent	—	—	—	Invention Steel
90	4.9	770	736	170	30	308	Absent	—	—	—	Comparative Steel
91	2.9	810	739	80	34	232	Absent	—	—	—	Comparative Steel
6	3.5	800	718	210	48	186	Absent	—	—	—	Comparative Steel
7	2.9	750	718	250	41	385	Absent	—	—	—	Invention Steel
8	4.3	820	718	200	41	443	Absent	—	—	—	Comparative Steel
10	4.2	760	718	160	50	411	Absent	—	—	—	Comparative Steel
11	2.9	800	718	110	26	280	Absent	—	—	—	Comparative Steel
12	2.9	750	718	240	36	434	Absent	—	—	—	Comparative Steel
13	3.3	770	718	200	37	484	Absent	—	—	—	Comparative Steel
14	3.4	820	718	190	33	450	Absent	—	—	—	Comparative Steel
15	4.4	780	718	90	33	340	Absent	—	—	—	Invention Steel
16	4.5	810	718	140	49	288	Absent	—	—	—	Comparative Steel
119	4.5	810	718	140	49	288	Absent	—	—	—	Comparative Steel
17	<u>20.0</u>	790	718	220	34	475	Absent	—	—	—	Comparative Steel
122	<u>30.0</u>	870	715	110	26	170	Absent	—	—	—	Comparative Steel
19	4.1	<u>1100</u>	718	250	55	223	Absent	—	—	—	Comparative Steel
20	4.5	860	718	230	31	154	Absent	—	—	—	Invention Steel

TABLE 2-2

Treatment No.	Kind of steel	Casting step: Average cooling rate at 950° C. to 550° C. in cooling of slab ° C./h	Casting step: Pressure applied at 950° C. to 550° C. N/cm^2	Heating step: Average heating rate in temperature range of 650° C. to 850° C. ° C./min	Heating step: Heating temperature ° C.	Hot rolling step: Cumulative rolling reduction in temperature range of 1050° C. or higher %	Cooling step: Time until start of rapid cooling after completion of hot railing step sec	Cooling step: Average cooling rate from start of cooling to 700° C. ° C./s	Coiling step Coiling temperature ° C.	Cold rolling step: Cold rolling reduction %	Cold rolling step: Sheet thickness after cold rolling mm	Leveling step Presence or absence of leveling
21	A	155	35	35	1240	41	2	90	480	52	1.20	Present
23	A	156	35	35	1240	40	2	98	550	52	1.20	Present
24	A	159	35	35	1240	45	1	93	500	52	1.20	Present
25	A	156	35	35	1240	44	1	90	500	52	1.20	Present
94	I	160	35	35	1240	40	2	90	530	52	1.20	Present
95	J	159	35	35	1240	40	1	96	560	52	1.20	Present
26	A	155	35	35	1240	41	1	94	470	52	1.20	Present
27	B	159	35	35	1240	43	1	91	530	52	1.20	Present
28	A	157	35	35	1240	43	1	98	540	70	0.50	Present
29	A	157	35	35	1240	43	2	90	550	52	1.20	Present
30	E	156	35	35	1240	42	1	94	500	52	1.20	Present
31	A	157	35	35	1240	40	2	100	540	52	1.20	Present
32	A	162	35	35	1240	42	1	92	490	52	1.20	Present
33	A	157	35	35	1240	45	1	93	500	52	1.20	Present
123	D	164	35	35	1240	43	1	94	530	52	1.20	Present
40	F	159	35	35	1240	42	2	100	570	52	1.20	Present
47	G	160	35	35	1240	42	2	95	540	52	1.20	Present
48	H	157	35	35	1240	44	1	94	480	52	1.20	Present
62	B	161	35	35	1240	42	1	100	570	52	0.80	Present
63	C	165	35	35	1240	43	2	91	530	52	1.20	Present
64	D	157	35	35	1240	41	2	97	510	52	1.20	Present
77	I	162	35	35	1240	41	2	98	530	52	1.20	Present
79	B	160	35	35	1240	45	2	90	560	52	1.20	Present
80	B	156	35	35	1240	40	2	94	560	52	1.20	Present
81	C	163	35	35	1240	42	1	99	540	52	1.20	Present
82	C	161	35	35	1240	44	1	96	560	52	1.20	Present

Treatment No.	Annealing step: Average heating rate during heating to Ac1 or higher ° C./s	Annealing step: Annealing temperature ° C.	Annealing step: AC1 ° C.	Annealing step: Retention time sec	Post-annealing cooling step Average cooling rate from annealing temperature to 550° C. or lower ° C./s	Retaining step Retention time in temperature range of 150° C. to 550° C. sec	Hot-dip galvanizing step: Presence or absence of plating treatment	Hot-dip galvanizing step: Kind of plating	Alloying step: Reheating (alloying) temperature ° C.	Alloying step: Retention time sec	Note
21	5.8	820	718	1	51	289	Absent	—	—	—	Comparative Steel
23	4.0	820	718	140	0.5	236	Absent	—	—	—	Comparative Steel
24	3.7	760	718	120	3	314	Absent	—	—	—	Invention Steel
25	5.1	790	718	160	10	348	Present	GI	—	—	Invention Steel
94	3.9	760	716	130	37	359	Absent	—	—	—	Invention Steel
95	5.1	740	718	160	47	400	Absent	—	—	—	Invention Steel
26	5.7	780	718	210	48	10	Absent	—	—	—	Comparative Steel
27	5.2	820	738	230	23	20	Absent	—	—	—	Invention Steel
28	4.3	820	718	150	28	800	Absent	—	—	—	Invention Steel
29	4.7	760	718	140	32	1200	Absent	—	—	—	Comparative Steel

TABLE 2-2-continued

30	4.3	760	718	160	34	312	Present	GI	—	—	Invention Steel
31	4.0	820	718	80	35	366	Present	GI	—	—	Invention Steel
32	4.5	810	718	190	25	386	Present	GA	480	28	Invention Steel
33	4.5	750	718	90	50	235	Present	GA	545	13	Invention Steel
123	4.5	820	752	220	55	290	Absent	—	—	—	Invention Steel
40	3.2	770	746	190	28	374	Absent	—	—	—	Invention Steel
47	5.6	750	736	180	20	314	Absent	—	—	—	Invention Steel
48	3.6	800	739	150	29	199	Absent	—	—	—	Invention Steel
62	3.6	750	738	240	42	332	Absent	—	—	—	Invention Steel
63	5.3	770	715	190	24	210	Absent	—	—	—	Invention Steel
64	4.9	810	752	230	22	379	Absent	—	—	—	Invention Steel
77	5.9	810	716	160	50	443	Absent	—	—	—	Invention Steel
79	4.3	760	738	200	41	242	Absent	—	—	—	Invention Steel
80	4.4	800	738	200	28	488	Absent	—	—	—	Invention Steel
81	2.9	740	715	120	36	237	Absent	—	—	—	Invention Steel
82	4.0	790	715	120	36	205	Absent	—	—	—	Invention Steel

TABLE 2-3

		Casting step		Heating step		Hot rolling step	Cooling step			Cold rolling step		Leveling step
Treat-ment No.	Kind of steel	Average cooling rate at 950° C. to 550° C. in cooling of slab ° C./h	Pressure applied at 950° C. to 550° C. N/cm^2	Average heating rate in temper-ature range of 650° C. to 850° C. ° C./min	Heating temper-ature ° C.	Cumulative rolling reduction in temper-ature range of 1050° C. or higher %	Time until start of rapid cooling after completion of hot railing step sec	Average cooling rate from start of cooling to 700° C. ° C./s	Coiling step Coiling temper-ature ° C.	Cold rolling reduction %	Sheet thickness after cold rolling mm	Presence or absence of leveling
83	D	161	35	35	1240	41	1	96	570	52	1.20	Present
84	D	155	35	35	1240	45	1	92	490	52	1.20	Present
85	E	159	35	35	1240	42	2	90	550	52	1.20	Present
124	E	164	35	35	1240	43	2	98	540	52	1.20	Present
125	F	157	35	35	1240	40	2	98	560	52	1.20	Present
88	F	159	35	35	1240	45	2	100	540	52	1.20	Present
89	G	165	35	35	1240	41	1	92	550	52	1.20	Present
126	G	161	35	35	1240	42	1	99	470	52	1.20	Present
127	H	161	35	35	1240	40	2	99	530	52	1.20	Present
92	H	164	35	35	1240	43	1	93	540	52	1.20	Present
93	I	156	35	35	1240	44	1	90	490	52	1.20	Present
128	I	160	35	35	1240	40	2	90	530	52	1.20	Present
129	J	159	35	35	1240	40	1	96	560	52	1.20	Present
96	K	155	35	35	1240	45	1	92	500	52	1.20	Present
97	L	160	35	35	1240	40	2	100	560	52	1.20	Present
98	M	161	35	35	1240	40	1	96	570	52	1.20	Present
99	N	158	35	35	1240	41	1	92	470	52	1.20	Present
100	O	158	35	35	1240	42	1	94	480	52	1.20	Present
101	P	160	35	35	1240	41	1	95	500	36	1.60	Present
102	Q	160	35	35	1240	44	2	91	570	50	2.00	Present
103	Q	163	35	35	1240	43	1	96	570	52	1.20	Present
104	R	160	35	35	1240	44	2	97	480	52	1.20	Present

TABLE 2-3-continued

105	R	165	35	35	1240	41	1	97	530	52	1.20	Present
106	S	165	35	35	1240	45	1	99	500	52	1.20	Present
107	S	155	35	35	1240	45	2	99	530	52	1.20	Present
108	T	160	35	35	1240	43	2	96	500	52	1.20	Present

	Annealing step				Post-annealing cooling step	Retaining step	Hot-dip galvanizing step		Alloying step		
Treatment No.	Average heating rate during heating to Ac1 or higher ° C./s	Annealing temperature ° C.	AC1 ° C.	Retention time sec	Average cooling rate from annealing temperature to 550° C. or lower ° C./s	Retention time in temperature range of 150° C. to 550° C. sec	Presence or absence of plating treatment	Kind of plating	Reheating (alloying) temperature ° C.	Retention time sec	Note
83	6.0	760	752	100	31	214	Absent	—	—	—	Invention Steel
84	5.6	800	752	180	34	301	Absent	—	—	—	Invention Steel
85	3.9	820	718	170	41	476	Absent	—	—	—	Invention Steel
124	5.4	820	718	160	54	276	Absent	—	—	—	Invention Steel
125	4.1	780	746	230	53	321	Absent	—	—	—	Invention Steel
88	3.4	760	746	80	36	225	Absent	—	—	—	Invention Steel
89	5.0	760	736	190	29	185	Absent	—	—	—	Invention Steel
126	5.7	770	736	170	30	308	Absent	—	—	—	Invention Steel
127	2.9	810	739	80	34	232	Absent	—	—	—	Invention Steel
92	3.8	770	739	120	36	163	Absent	—	—	—	Invention Steel
93	4.3	750	716	170	49	162	Absent	—	—	—	Invention Steel
128	3.9	760	716	130	37	359	Absent	—	—	—	Invention Steel
129	5.1	740	718	160	47	400	Absent	—	—	—	Invention Steel
96	4.3	770	733	210	45	202	Absent	—	—	—	Invention Steel
97	4.4	740	715	90	52	407	Absent	—	—	—	Invention Steel
98	5.7	788	757	140	49	498	Absent	—	—	—	Invention Steel
99	4.1	800	719	150	48	463	Absent	—	—	—	Invention Steel
100	3.0	760	745	190	50	299	Absent	—	—	—	Invention Steel
101	3.1	790	717	90	28	367	Absent	—	—	—	Invention Steel
102	5.9	820	718	190	39	174	Absent	—	—	—	Invention Steel
103	4.3	790	718	180	22	346	Absent	—	—	—	Invention Steel
104	5.9	800	739	140	51	270	Absent	—	—	—	Invention Steel
105	3.6	750	739	190	35	198	Absent	—	—	—	Invention Steel
106	3.7	790	715	200	46	233	Absent	—	—	—	Invention Steel
107	6.0	810	715	170	40	413	Absent	—	—	—	Invention Steel
108	3.2	820	752	130	28	205	Absent	—	—	—	Invention Steel

TABLE 2-4

Treatment No.	Kind of steel	Casting step: Average cooling rate at 950° C. to 550° C. in cooling of slab ° C./h	Casting step: Pressure applied at 950° C. to 550° C. N/cm^2	Heating step: Average heating rate in temperature range of 650° C. to 850° C. ° C./min	Heating step: Heating temperature ° C.	Hot rolling step: Cumulative rolling reduction in temperature range of 1050° C. or higher %	Cooling step: Time until start of rapid cooling after completion of hot railing step sec	Cooling step: Average cooling rate from start of cooling to 700° C. ° C./s	Coiling step: Coiling temperature ° C.	Cold rolling step: Cold rolling reduction %	Cold rolling step: Sheet thickness after cold rolling mm	Leveling step: Presence or absence of leveling
109	T	162	35	35	1240	42	2	100	480	52	1.20	Present
130	A	158	35	35	1240	41	2	96	490	52	1.20	Absent
120	B	161	35	35	1240	42	1	100	570	52	1.20	Absent
121	C	163	35	35	1240	42	1	99	540	52	1.20	Absent
110	aa	155	35	35	1240	44	2	98	500	52	1.20	Present
111	bb	156	35	35	1240	42	1	98	470	52	1.20	Present
112	cc	158	35	35	1240	43	2	92	570	52	1.20	Present
113	dd	156	35	35	1240	42	2	95	540	52	1.20	Present
114	ee	160	35	35	1240	45	1	94	550	52	1.20	Present
115	ff	165	35	35	1240	44	2	98	520	52	1.20	Present
116	gg	160	35	35	1240	41	1	97	540	52	1.20	Present
117	hh	165	35	35	1240	40	2	90	510	52	1.20	Present
118	ii	155	35	35	1240	41	1	99	490	52	1.20	Present
a01	J	108	17	25	1225	70	1	68	580	75	0.50	Present
a02	L	326	25	28	1240	55	1	70	535	40	1.60	Present
a03	N	203	67	15	1280	80	1	69	540	60	0.80	Present
a04	P	116	12	18	1280	70	2	59	515	70	1.20	Present
a05	R	278	44	46	1265	75	1	73	550	80	0.40	Present
a06	T	125	15	8	1185	60	2	65	620	50	1.00	Present
a07	U	223	24	41	1195	60	1	68	630	75	0.40	Present
a08	V	216	23	19	1195	85	1	90	470	60	0.80	Present
a09	W	278	16	28	1260	76	2	58	520	70	0.45	Present
a10	X	137	56	24	1185	79	2	64	565	50	0.80	Present
a11	Y	158	37	27	1225	63	1	101	575	50	2.00	Present
a12	Z	236	31	40	1245	58	2	71	605	75	0.50	Present

Treatment No.	Annealing step: Average heating rate during heating to Ac1 or higher ° C./s	Annealing step: Annealing temperature ° C.	Annealing step: AC1 ° C.	Annealing step: Retention time sec	Post-annealing cooling step: Average cooling rate from annealing temperature to 550° C. or lower ° C./s	Retaining step: Retention time in temperature range of 150° C. to 550° C. sec	Hot-dip galvanizing step: Presence or absence of plating treatment	Hot-dip galvanizing step: Kind of plating	Alloying step: Reheating (alloying) temperature ° C.	Alloying step: Retention time sec	Note
109	4.8	780	752	130	24	225	Absent	—	—	—	Invention Steel
130	4.8	750	718	120	30	285	Absent	—	—	—	Invention Steel
120	3.8	760	738	250	45	328	Absent	—	—	—	Invention Steel
121	3.3	750	715	130	38	230	Absent	—	—	—	Invention Steel
110	4.9	790	715	80	27	355	Absent	—	—	—	Comparative Steel
111	3.3	810	713	200	34	362	Absent	—	—	—	Comparative Steel
112	3.1	810	775	120	46	382	Absent	—	—	—	Comparative Steel
113	5.7	750	677	180	31	385	Absent	—	—	—	Comparative Steel
114	4.7	780	724	160	36	178	Absent	—	—	—	Comparative Steel
115	3.8	780	694	230	37	407	Absent	—	—	—	Comparative Steel

TABLE 2-4-continued

116	3.5	820	705	250	48	304	Absent	—	—	—	Comparative Steel
117	5.8	740	703	210	55	281	Absent	—	—	—	Comparative Steel
118	5.5	770	708	130	54	454	Absent	—	—	—	Comparative Steel
a01	3.3	765	718	100	35	87	Present	GA	540	20	Invention Steel
a02	3.0	780	715	75	20	270	Present	GI	—	—	Invention Steel
a03	1.8	790	719	120	23	63	Absent	—	—	—	Invention Steel
a04	2.3	800	717	90	33	107	Present	GA	535	16	Invention Steel
a05	3.9	785	739	125	65	45	Absent	—	—	—	Invention Steel
a06	2.5	810	752	100	25	105	Present	GI	—	—	Invention Steel
a07	2.2	765	720	40	29	79	Present	GA	501	47	Invention Steel
a08	2.1	790	700	80	47	58	Absent	—	—	—	Invention Steel
a09	4.5	785	709	150	16	49	Present	GI	—	—	Invention Steel
a10	3.5	805	723	60	69	630	Absent	—	—	—	Invention Steel
a11	3.7	765	710	170	25	258	Present	GA	495	33	Invention Steel
a12	1.6	790	706	80	29	438	Absent	—	—	—	Invention Steel

The obtained steel sheet was evaluated by observing the microstructure and measuring the mechanical properties.

In the observation of the microstructure, a sample with a sheet thickness direction cross section parallel to the rolling direction as an observed section was collected, and in a range between a ⅛ thickness and a ⅜ thickness with a ¼ thickness as the center, the area fraction of each structure was measured by the above-described method.

In addition, the average distance between the centers of high Mn regions at the ¼ thickness in a plane parallel to the rolling direction of the steel sheet and a plane perpendicular to the sheet thickness direction, the ratio (D_A/D_B) between the density D_A of the high Mn regions at a sheet width center portion and the density D_B of the high Mn regions at a ¼ width, the ratio of the average hardness of the high Mn regions to the average hardness of low Mn regions, and the difference between the average value of a top 5% and the average value of a bottom 5% of Mn contents in the low Mn regions were obtained by the above-described method.

A tensile test was conducted according to JIS Z 2241: 2011, and the mechanical properties (yield stress, tensile strength, and elongation) were evaluated. The measured position of the sheet was the sheet width center portion, and the test direction was a direction perpendicular to the rolling direction. The shape of the test piece was a No. 5 test piece shown in JIS Z 2241:2011.

Shape-fixability is improved as long as the yield stress can be reduced while increasing the tensile strength. Therefore, the shape-fixability was evaluated by the tensile strength and the yield ratio (YP/TS). In a case where the tensile strength TS was 590 MPa or more and YP/TS≤0.80, excellent shape-fixability was determined.

when the ductility decreases, there is a possibility that press forming itself may not be possible. Therefore, workability was evaluated by the product of the tensile strength and elongation (TS×EL). A case of TS×EL≥14,000 MPa·% was determined to have sufficient workability.

Dimensional precision after pressing, which is an object of the present invention, is improved by reducing the difference in work hardening depending on the location. Therefore, in the present invention, the dimensional precision after pressing was evaluated by the difference in the amount of work hardening depending on the location.

The difference in the amount of work hardening depending on the location was defined as follows.

First, the true stress is indicated as σ, the true strain is indicated as E, and the true stress is differentiated by the true strain to obtain dσ/dε. Then, a graph of dσ/dε and σ is drawn. When the yield stress at the true stress is indicated as σYP, dσ/dε has almost the same value when σ is 0 to σYP. Thereafter, dσ/dε decreases. The slope (the graph of dσ/dε and σ) has an inflection point in the middle and becomes gentle from the middle. The true stress at the inflection point is defined as σin. This is because the amount of work hardening at σin may vary greatly depending on the location of the steel sheet.

In addition, dσ/dε (σin) at a sheet width center portion and dσ/dε (σin) at a ¼ width portion are obtained, and the absolute value of the difference therebetween is defined as |Δdσ/dε (σin)|W and obtained. Similarly, in order to investigate the difference in the amount of work hardening in the rolling direction at the sheet width center portion, when any position of the sheet width center portion is defined as a sheet width center portion 1 and the position of the sheet width center portion at a position 500 to 1000 mm away therefrom is defined as a sheet width center portion 2, the absolute value of the difference in dσ/dε (σin) is defined as |Δdσ/dε (σin)|L. The sheet width center portion 1 is the same as the sheet width center portion used for |Δdσ/dε (σin)|W.

In the present example, the difference in the amount of work hardening depending on the location was evaluated by |Δdσ/dε (σin)|L and |Δdσ/dε (σin)|W.

In a case of |Δdσ/dε (σin)|L≤1500 and |Δdσ/dε (σin)|W≤1500, it was determined that the difference in work hardening depending on the location was small.

Tables 3-1 to 3-4 show the measurement results and evaluation results.

The chemical composition of each of the obtained steel sheets was substantially the same as the chemical composition of the corresponding molten steel.

TABLE 3-1

Treatment No.	Ferrite fraction area %	Fraction of residual austenite and martensite area %	Bainite fraction area %	Pearlite fraction area %	Average distance between centers of high Mn regions in plane at 1/4 thickness mm	Density D_A of high Mn regions at sheet width center portion /mm^2	Density D_B of high Mn regions at 1/4 width /mm^2	D_A/D_B —	Average hardness of high Mn regions HV	Average hardness of low Mn regions HV	Ratio of average hardness of high Mn regions to average hardness of low Mn regions —	Difference between average value of top 5% and average value of bottom 5% of Mn contents in low Mn regions mass %
1	78	13	9	0	0.60	2.35	2.08	1.13	321	204	1.57	0.42
18	86	8	6	0	0.53	1.91	1.32	1.45	221	95	2.34	0.41
37	59	7	34	0	0.59	1.27	1.80	1.42	213	88	2.43	0.44
2	79	12	9	0	0.75	2.71	1.87	1.45	419	165	2.53	0.48
3	78	13	9	0	0.74	2.42	1.94	1.25	320	163	1.96	0.48
86	78	13	9	0	1.52	2.41	1.71	1.41	397	171	2.32	0.41
87	16	24	60	0	1.62	1.49	2.11	0.71	570	239	2.38	0.45
4	79	13	8	0	1.72	2.52	1.75	1.44	374	161	2.32	0.50
5	78	14	8	0	0.90	2.43	1.94	1.25	319	168	1.90	0.47
90	29	22	49	0	0.50	1.84	1.76	1.05	367	334	1.10	0.12
91	30	18	52	0	0.48	2.09	1.91	1.09	367	306	1.20	0.11
6	77	15	8	0	0.47	2.18	1.98	1.10	298	271	1.10	0.13
7	79	13	8	0	0.45	2.23	1.93	1.15	294	268	1.10	0.35
8	91	2	7	0	0.63	1.83	2.08	0.88	260	166	1.57	0.47
10	76	15	9	0	0.53	2.41	1.75	1.38	397	167	2.38	0.50
11	78	13	9	0	1.23	2.37	1.79	1.32	398	171	2.33	0.47
12	79	12	9	0	1.22	2.37	1.79	1.33	393	169	2.32	0.48
13	79	12	9	0	1.24	2.37	1.79	1.33	395	170	2.33	0.45
14	78	13	9	0	1.22	2.38	1.78	1.34	399	172	2.32	0.43
15	77	14	9	0	0.64	2.21	1.95	1.11	290	181	1.60	0.48
16	77	14	9	0	1.20	2.43	1.73	1.38	392	178	2.22	0.49
119	77	14	9	0	0.61	2.43	1.73	1.12	392	178	1.20	0.11
17	78	13	9	0	0.78	2.40	1.76	1.10	390	177	1.58	0.11
122	87	7	6	0	0.77	2.40	1.76	1.04	390	177	1.52	0.13
19	0	4	96	0	0.69	1.84	2.08	0.88	340	203	1.68	0.45
20	30	20	50	0	0.56	2.18	1.90	1.15	286	194	1.48	0.46

Treatment No.	Yield stress MPa	Tensile strength MPa	Yield ratio —	Elongation %	TS × EL MPa-%	dσ/dε (σin) in sheet width center portion 1 MPa	dσ/dε (σin) in sheet width center portion 2 MPa	dσ/dε (σin) in 1/4 width portion MPa	|Δdσ/dε (σin)|L MPa	|Δdσ/dε (σin)|W MPa	Note —
1	497	841	0.59	26	21450	12504	13167	11864	663	640	Invention Steel
18	333	650	0.51	30	19500	10224	10696	8332	472	1892	Comparative Steel
37	319	632	0.51	36	23000	7353	7001	9142	352	1789	Comparative Steel
2	501	845	0.59	25	20800	12552	13128	10542	576	2010	Comparative Steel
3	503	847	0.59	25	20800	12576	11928	11373	648	1203	Invention Steel
86	507	852	0.60	21	17550	13136	12517	11445	619	1691	Comparative Steel
87	903	1214	0.74	18	21467	5230	5480	6749	250	1519	Comparative Steel
4	497	841	0.59	22	18850	12504	11871	10554	633	1950	Comparative Steel
5	499	843	0.59	22	18850	12528	11920	11330	608	1198	Invention Steel
90	631	1052	0.60	20	21467	6640	5125	6341	1515	299	Comparative Steel
91	616	1010	0.61	20	20700	6850	5321	6477	1529	373	Comparative Steel
6	508	853	0.59	21	19500	12648	10787	13309	1861	661	Comparative Steel
7	514	843	0.61	25	21450	12528	13871	11867	1343	661	Invention Steel
8	512	857	0.60	16	13980	12696	12102	13311	594	615	Comparative Steel
10	501	845	0.59	24	20150	13052	12389	11371	663	1682	Comparative Steel
11	508	853	0.61	21	20800	12148	12810	13780	662	1632	Comparative Steel
12	499	843	0.59	22	18850	13028	13689	11374	661	1654	Comparative Steel
13	503	847	0.61	25	21450	13076	13726	11435	650	1641	Comparative Steel
14	512	857	0.60	20	20800	13196	13840	11551	644	1644	Comparative Steel
15	526	849	0.62	24	20150	11600	11314	11030	286	559	Invention Steel
16	510	855	0.83	24	17550	12672	12051	14480	620	1808	Comparative Steel
119	510	854	0.83	24	20150	12674	14478	12051	1804	623	Comparative Steel
17	506	851	0.84	25	18200	12624	10808	13263	1816	639	Comparative Steel
122	333	650	0.82	30	21450	10224	8500	9753	1723	471	Comparative Steel
19	790	1120	0.71	12	13306	15836	15112	16584	724	748	Comparative Steel
20	520	865	0.60	23	19809	12791	12150	12127	641	664	Invention Steel

TABLE 3-2

Treatment No.	Ferrite fraction area %	Fraction of residual austenite and martensite area %	Bainite fraction area %	Pearlite fraction area %	Average distance between centers of high Mn regions in plane at ¼ thickness mm	Density D_A of high Mn regions at sheet width center portion /mm^2	Density D_B of high Mn regions at ¼ width /mm^2	D_A/D_B —	Average hardness of high Mn regions HV	Average hardness of low Mn regions HV	Ratio of average hardness of high Mn regions to average hardness of low Mn regions —	Difference between average value of top 5% and average value of bottom 5% of Mn contents in low Mn regions mass %
21	93	0	7	0	0.53	2.16	1.91	1.13	228	168	1.36	0.45
23	67	8	6	19	0.51	1.81	2.09	0.87	204	154	1.33	0.47
24	75	12	7	6	0.56	1.84	2.08	0.89	235	168	1.40	0.42
25	76	15	9	0	0.57	2.35	2.07	1.14	325	220	1.48	0.46
94	89	5	6	0	0.70	1.84	1.60	1.15	200	120	1.66	0.45
95	81	12	7	0	0.57	2.31	2.06	1.12	293	224	1.31	0.48
26	74	16	10	0	0.77	2.22	1.94	1.14	390	177	2.10	0.46
27	51	15	34	0	0.72	1.71	1.93	0.88	390	177	1.98	0.45
28	68	13	19	0	0.63	1.82	2.07	0.88	257	179	1.44	0.48
29	74	2	24	0	0.52	1.97	2.25	0.88	298	216	1.38	0.45
30	79	13	8	0	0.50	2.15	1.87	1.15	277	210	1.32	0.44
31	76	15	9	0	0.57	2.15	1.90	1.13	277	195	1.42	0.46
32	79	12	9	0	0.64	1.84	2.08	0.88	259	157	1.65	0.41
33	80	11	9	0	0.55	2.16	1.91	1.14	276	206	1.34	0.43
123	60	6	34	0	0.51	1.40	1.59	0.88	190	140	1.36	0.48
40	20	24	56	0	0.58	1.73	1.95	0.88	429	270	1.59	0.48
47	20	22	58	0	0.62	2.02	1.81	1.11	394	272	1.45	0.48
48	34	17	49	0	0.52	2.07	1.84	1.13	327	244	1.34	0.43
62	47	11	42	0	0.51	1.91	1.66	1.15	266	189	1.41	0.45
63	87	7	6	0	0.63	1.87	1.64	1.14	210	131	1.61	0.47
64	60	7	33	0	0.62	1.55	1.73	0.90	221	137	1.62	0.49
77	87	6	7	0	0.62	1.53	1.76	0.87	186	119	1.57	0.48
79	45	11	44	0	0.60	1.91	1.66	1.15	266	160	1.66	0.43
80	45	13	42	0	0.59	2.05	1.82	1.12	305	189	1.62	0.47
81	87	7	6	0	0.53	1.58	1.79	0.88	191	137	1.40	0.46
82	87	7	6	0	0.57	2.03	1.80	1.13	246	165	1.49	0.48

Treatment No.	Yield stress MPa	Tensile strength MPa	Yield ratio —	Elongation %	TS × EL MPa-%	dσ/dε (σin) in sheet width center portion 1 MPa	dσ/dε (σin) in sheet width center portion 2 MPa	dσ/dε (σin) in ¼ width portion MPa	|Δdσ/dε (σin)|L MPa	|Δdσ/dε (σin)|W MPa	Note —
21	372	699	0.53	19	13082	10809	10274	10298	535	511	Comparative Steel
23	346	666	0.52	19	12689	10415	9924	10981	491	566	Comparative Steel
24	439	777	0.56	19	14890	11740	11196	12278	545	538	Invention Steel
25	506	851	0.60	21	17550	12624	13212	11935	588	689	Invention Steel
94	295	600	0.49	36	21450	9627	9183	9127	444	500	Invention Steel
95	438	776	0.56	27	20800	11728	12323	11155	595	573	Invention Steel
26	505	849	0.59	25	21450	13200	12569	11338	631	1862	Comparative Steel
27	470	812	0.58	30	24533	7440	7092	8633	348	1193	Invention Steel
28	643	857	0.75	26	22231	12696	12017	13298	679	602	Invention Steel
29	693	845	0.82	25	21234	12552	13133	13241	581	689	Comparative Steel
30	496	840	0.59	22	18850	12493	11814	11839	678	654	Invention Steel
31	512	857	0.60	25	21450	12696	11999	12108	696	587	Invention Steel
32	510	855	0.60	21	17690	12672	12080	13268	592	596	Invention Steel
33	499	843	0.59	22	18209	12528	11918	11928	611	601	Invention Steel

TABLE 3-2-continued

123	319	632	0.51	40	25300	8740	8286	9160	454	420	Invention Steel
40	901	1212	0.74	18	22233	5840	6152	6143	312	303	Invention Steel
47	708	1048	0.68	20	21467	6660	7023	6356	363	304	Invention Steel
48	665	1008	0.66	21	20700	6860	6519	6551	341	309	Invention Steel
62	458	798	0.57	31	24533	7910	7518	7480	392	430	Invention Steel
63	327	642	0.51	29	18850	10128	9606	9630	523	499	Invention Steel
64	318	630	0.50	34	21467	8750	9210	9158	460	408	Invention Steel
77	303	610	0.50	33	20150	9746	9292	10245	455	499	Invention Steel
79	459	800	0.57	26	20700	7900	7512	7483	388	417	Invention Steel
80	467	808	0.58	28	23000	7860	8250	7471	390	389	Invention Steel
81	322	636	0.51	28	17550	10057	9534	10540	522	483	Invention Steel
82	330	646	0.51	29	18850	10176	10710	9661	534	515	Invention Steel

TABLE 3-3

Treatment No.	Ferrite fraction area %	Fraction of residual austenite and martensite area %	Bainite fraction area %	Pearlite fraction area %	Average distance between centers of high Mn regions in plane at ¼ thickness mm	Density D_A of high Mn regions at sheet width center portion /mm^2	Density D_B of high Mn regions at ¼ width /mm^2	D_A/D_B —	Average hardness of high Mn regions HV	Average hardness of low Mn regions HV	Ratio of average hardness of high Mn regions to average hardness of low Mn regions —	Difference between average value of top 5% and average value of bottom 5% of Mn contents in low Mn regions mass %
83	64	7	29	0	0.69	1.67	1.46	1.14	204	121	1.69	0.49
84	60	7	33	0	0.61	1.66	1.46	1.14	206	142	1.45	0.49
85	77	15	8	0	0.57	2.15	1.88	1.14	280	215	1.30	0.41
124	78	14	8	0	0.51	2.15	1.88	1.14	280	204	1.37	0.45
125	17	17	66	0	0.61	1.73	1.95	0.89	428	304	1.41	0.49
88	16	24	60	0	0.59	1.87	1.65	1.13	396	259	1.53	0.42
89	25	19	56	0	0.58	1.73	1.94	0.89	366	265	1.38	0.43
126	31	20	49	0	0.58	1.87	1.66	1.13	344	246	1.40	0.48
127	23	21	56	0	0.53	2.08	1.84	1.13	331	245	1.35	0.42
92	32	19	49	0	0.58	2.07	1.82	1.14	325	237	1.37	0.42
93	89	5	6	0	0.51	1.82	1.60	1.14	197	151	1.30	0.49
128	89	5	6	0	0.58	1.96	1.76	1.12	226	163	1.39	0.43
129	79	12	9	0	0.64	1.97	2.22	0.89	271	175	1.55	0.42
96	47	12	41	0	0.64	2.03	1.82	1.12	293	189	1.55	0.45
97	85	8	7	0	0.60	1.86	1.65	1.13	211	150	1.41	0.43
98	61	7	32	0	0.54	1.35	1.21	1.12	227	150	1.51	0.46
99	79	13	8	0	0.55	2.14	1.89	1.13	258	176	1.47	0.41
100	11	24	65	0	0.65	2.00	1.74	1.15	394	246	1.60	0.41
101	74	17	9	0	0.54	1.87	1.65	1.13	321	221	1.45	0.44
102	79	13	8	0	0.51	1.82	2.06	0.88	239	184	1.30	0.49
103	82	11	7	0	0.61	2.17	1.89	1.15	263	189	1.39	0.46
104	51	11	38	0	0.61	2.00	1.75	1.14	260	175	1.49	0.48
105	54	11	35	0	0.62	1.98	1.75	1.14	254	172	1.48	0.43
106	85	9	6	0	0.54	1.89	1.64	1.15	198	143	1.38	0.48
107	88	6	6	0	0.63	2.01	1.80	1.12	225	157	1.43	0.49
108	62	10	28	0	0.61	1.79	1.60	1.11	225	145	1.55	0.41

TABLE 3-3-continued

Treatment No.	Yield stress MPa	Tensile strength MPa	Yield ratio —	Elongation %	TS × EL MPa-%	dσ/dε (σin) in sheet width center portion 1 MPa	dσ/dε (σin) in sheet width center portion 2 MPa	dσ/dε (σin) in ¼ width portion MPa	|Δdσ/dε (σin)|L MPa	|Δdσ/dε (σin)|W MPa	Note —
83	310	620	0.50	33	20700	8800	8362	8365	438	435	Invention Steel
84	316	628	0.50	34	21467	8760	8289	8321	471	439	Invention Steel
85	507	852	0.60	25	21450	12636	11994	12012	642	624	Invention Steel
124	507	852	0.60	25	21450	12636	11973	12000	663	636	Invention Steel
125	903	1214	0.74	14	17589	5830	6129	6122	299	292	Invention Steel
88	898	1210	0.74	15	17684	5850	5572	5575	278	275	Invention Steel
89	711	1050	0.68	23	23767	6650	6953	6967	303	317	Invention Steel
126	713	1052	0.68	20	20700	6640	6339	6335	301	305	Invention Steel
127	667	1010	0.66	20	20700	6850	6522	6524	328	326	Invention Steel
92	658	1002	0.66	24	23767	6890	6515	6560	375	330	Invention Steel
93	294	598	0.49	32	18850	9603	9114	9122	489	481	Invention Steel
128	295	600	0.49	29	17550	9627	10132	9171	505	456	Invention Steel
129	438	776	0.56	25	19500	11728	12260	12318	532	590	Invention Steel
96	443	782	0.57	27	21467	7990	8373	7616	383	374	Invention Steel
97	338	656	0.52	28	18200	10296	9739	9828	556	468	Invention Steel
98	292	596	0.49	37	22233	8922	9459	8504	537	418	Invention Steel
99	453	793	0.57	22	17550	11931	11322	11366	609	566	Invention Steel
100	861	1180	0.73	20	23767	6000	5717	5682	284	318	Invention Steel
101	641	986	0.65	22	21450	14236	13534	13580	702	656	Invention Steel
102	452	792	0.57	26	20800	11919	11331	12475	588	555	Invention Steel
103	447	786	0.57	22	17550	11848	11313	11219	535	629	Invention Steel
104	449	788	0.57	26	20700	7960	7570	7556	390	404	Invention Steel
105	440	778	0.57	28	21467	8010	7603	7626	407	384	Invention Steel
106	292	596	0.49	32	18850	9579	9106	9076	473	503	Invention Steel
107	295	600	0.49	35	20800	9627	10074	9168	447	459	Invention Steel
108	297	602	0.49	37	22233	8890	9331	8482	441	408	Invention Steel

TABLE 3-4

Treatment No.	Ferrite fraction area %	Fraction of residual austenite and martensite area %	Bainite fraction area %	Pearlite fraction area %	Average distance between centers of high Mn regions in plane at ¼ thickness mm	Density D_A of high Mn regions at sheet width center portion /mm^2	Density D_B of high Mn regions at ¼ width /mm^2	D_A/D_B —	Average hardness of high Mn regions HV	Average hardness of low Mn regions HV	Ratio of average hardness of high Mn regions to average hardness of low Mn regions —	Difference between average value of top 5% and average value of bottom 5% of Mn contents in low Mn regions mass %
109	63	8	29	0	0.62	1.55	1.74	0.89	211	132	1.60	0.42
130	72	20	8	0	0.65	2.34	1.82	1.28	331	215	1.54	0.43
120	42	17	41	0	0.58	2.05	1.59	1.29	376	200	1.88	0.46
121	82	13	5	0	0.59	2.01	1.59	1.27	201	151	1.33	0.48
110	88	6	6	0	0.67	1.68	1.92	0.88	279	118	2.37	0.43
111	98	0	2	0	0.60	3.16	2.79	1.13	156	110	1.42	0.46
112	51	13	36	0	0.51	3.53	3.07	1.15	368	159	2.31	0.47
113	80	12	8	0	0.50	4.80	4.20	1.14	364	155	2.35	0.47
114	95	0	5	0	0.49	0.05	0.05	1.15	153	118	1.32	0.48
115	81	11	8	0	0.55	3.09	2.71	1.14	268	192	1.43	0.48
116	81	11	8	0	0.58	1.92	1.68	1.14	272	182	1.54	0.47
117	80	12	8	0	0.56	1.88	2.12	0.89	364	153	2.38	0.50
118	79	12	9	0	0.59	1.51	1.69	0.89	364	157	2.32	0.42
a01	74	12	12	2	0.68	1.75	2.00	0.88	287	174	1.65	0.51
a02	68	13	15	4	0.43	2.85	2.65	1.08	228	149	1.53	0.48
a03	66	24	10	0	0.45	2.66	2.80	0.95	335	261	1.28	0.34
a04	70	15	13	2	0.60	2.01	1.74	1.16	272	195	1.39	0.53
a05	61	28	11	0	0.38	3.05	2.45	1.24	361	254	1.42	0.37
a06	70	8	19	3	0.73	1.36	1.50	0.91	232	135	1.72	0.58
a07	92	6	0	2	0.64	1.99	2.17	0.92	224	163	1.37	0.35
a08	83	12	5	0	0.47	2.70	2.37	1.14	253	191	1.32	0.51
a09	90	7	2	1	0.53	2.32	2.17	1.07	235	129	1.82	0.45
a10	68	4	25	3	0.40	2.96	3.02	0.98	216	175	1.23	0.32

TABLE 3-4-continued

a11	82	11	6	1	0.48	2.61	2.83	0.92	293	155	1.89	0.66
a12	87	9	4	0	0.50	2.25	2.38	0.95	260	183	1.42	0.61

Treatment No.	Yield stress MPa	Tensile strength MPa	Yield ratio —	Elongation %	TS × EL MPa-%	dσ/dε (σin) in sheet width center portion 1 MPa	dσ/dε (σin) in sheet width center portion 2 MPa	dσ/dε (σin) in ¼ width portion MPa	|Δdσ/dε (σin)|L MPa	|Δdσ/dε (σin)|W MPa	Note —
109	291	594	0.49	43	25300	8930	9451	9370	521	440	Invention Steel
130	508	851	0.59	25	21350	12609	12178	11288	431	1321	Invention Steel
120	468	802	0.57	29	23582	8021	7618	6693	403	1328	Invention Steel
121	333	651	0.51	26	16999	10150	9642	8835	508	1315	Invention Steel
110	292	596	0.49	35	13890	9079	9527	10586	448	1507	Comparative Steel
111	170	410	0.41	44	18200	7358	7713	6976	355	382	Comparative Steel
112	450	790	0.57	29	23000	8550	8108	6915	442	1635	Comparative Steel
113	440	778	0.57	23	18050	10652	11194	12198	542	1545	Comparative Steel
114	190	444	0.43	44	19500	7764	7363	7349	401	415	Comparative Steel
115	445	711	0.57	26	13209	11824	11269	11227	555	597	Comparative Steel
116	452	702	0.57	27	13940	11919	12505	11294	586	626	Comparative Steel
117	438	776	0.56	28	21450	11228	10620	12795	608	1566	Comparative Steel
118	443	782	0.57	27	20800	11300	10778	12841	522	1541	Comparative Steel
a01	517	807	0.64	22	17518	9277	8675	9002	602	275	Invention Steel
a02	408	712	0.57	24	17169	9750	9701	9080	49	670	Invention Steel
a03	617	997	0.62	15	14561	10592	10259	10663	333	71	Invention Steel
a04	547	867	0.63	21	18034	9991	10362	9698	371	293	Invention Steel
a05	484	1145	0.42	18	20948	11944	11165	12045	779	101	Invention Steel
a06	368	598	0.62	26	15316	10895	11403	10524	508	371	Invention Steel
a07	261	601	0.43	25	15205	8023	7558	8245	465	222	Invention Steel
a08	352	725	0.49	23	16670	10372	10008	10818	364	446	Invention Steel
a09	303	623	0.49	25	15393	7733	7333	7346	400	387	Invention Steel
a10	443	653	0.68	24	15943	6895	7023	6737	128	158	Invention Steel
a11	439	883	0.50	20	17315	11039	10265	10240	774	799	Invention Steel
a12	371	791	0.47	19	15189	10852	10138	10523	714	329	Invention Steel

In Treatment Nos. 2, 18, and 37, the average cooling rate at 950° C. to 550° C. was slow in the cooling of the slab, and Expression (1) and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions were outside the ranges of the present invention. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment Nos. 4, 86, and 87, the pressure applied at 950° C. to 550° C. was low, the average distance between the centers of the high Mn regions at the ¼ thickness, Expression (1), and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions were outside the ranges of the present invention. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment Nos. 6, 90, and 91, the heating rate in the temperature range of 650° C. to 850° C. in the heating before the hot rolling was fast, and the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents in the low Mn regions was small. As a result, |Δdσ/dε (σin)|L was outside the target range.

In Treatment No. 8, the heating temperature during the hot rolling was low, and the fraction of residual austenite and martensite was low. As a result, TS×EL was outside the target range.

In Treatment No. 10, the rolling reduction in the temperature range of 1050° C. or higher was low, D_A/D_B did not satisfy Expression (1), and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was high. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment Nos. 11 and 12, the time from the end of the hot rolling to the start of the cooling was long, D_A/D_B did not satisfy Expression (1), and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was high. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment Nos. 13 and 14, the average cooling rate from the start of the cooling to 700° C. was slow, D_A/D_B did not satisfy Expression (1), and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was high. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment No. 16, the coiling temperature was high, D_A/D_B did not satisfy Expression (1), and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was high, so that |Δdσ/dε (σin)|W was outside the target range.

In Treatment No. 119, the coiling temperature was low, and the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents in the low Mn regions was small. As a result, |Δdσ/dε (σin)|L was outside the target range.

In Treatment Nos. 17 and 122, the average heating rate during the heating to Ac1° C. or higher was fast, and the difference between the average value of the top 5% and the average value of the bottom 5% of the Mn contents in the low Mn regions was small. As a result, |Δdσ/dε (σin)|L was outside the target range.

In Treatment No. 19, the annealing temperature was high, and the ferrite fraction was outside the range of the present invention. As a result, TS×EL was outside the target range.

In Treatment No. 21, the retention time at the annealing temperature was short, and the fraction of residual austenite and martensite was low. As a result, TS×EL was outside the target range.

In Treatment No. 23, the average cooling rate from the annealing temperature to 550° C. was slow, and the pearlite fraction was higher than the range of the present invention. As a result, TS×EL was outside the target range.

In Treatment No. 26, the retention time in the temperature range of 150° C. to 550° C. was short, and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was higher than the range of the present invention. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment No. 29, the retention time in the temperature range of 150° C. to 550° C. was long, and residual austenite and martensite were low. As a result, the yield ratio was outside the target range.

In Treatment No. 110, since Kind of steel aa was used, the Nb content was higher than the range of the present invention, and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was higher than the range of the present invention. As a result, TS×EL was low and |Δdσ/dε (σin)|W was outside the target range.

In Treatment No. 111, since Kind of steel bb was used, the C content was lower than the range of the present invention, so that the ferrite fraction was high and the residual austenite fraction was low. As a result, TS was outside the target range.

In Treatment No. 112, since Kind of steel cc was used, the Si content was higher than the range of the present invention, and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was higher than the range of the present invention. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment No. 113, since Kind of steel dd was used, the Mn content was higher than the range of the present invention, and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was higher than the range of the present invention. As a result, |Δdσ/dε (σin)|W was outside the target range.

In Treatment No. 114, since Kind of steel ee was used, the Mn content was lower than the range of the present invention, and residual austenite and martensite were lower than the ranges of the present invention. As a result, TS was outside the target range.

In Treatment No. 115, since Kind of steel ff was used, the P content was higher than the range of the present invention. As a result, TS×EL was outside the target range.

In Treatment No. 116, since Kind of steel gg was used, the S content was higher than the range of the present invention. As a result, TS×EL was outside the target range.

In Treatment No. 117, since Kind of steel hh was used, the Al content was higher than the range of the present invention. As a result, the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was higher than the range of the present invention, and |Δdσ/dε (σin)|W was outside the target range.

Treatment No. 118 is a component of symbol ii, the Ti content was higher than the range of the present invention, and the ratio of the average hardness of the high Mn regions to the average hardness of the low Mn regions was higher than the range of the present invention. As a result, |Δdσ/dε (σin)|W was outside the target range.

Regarding the other conditions, the structure was within the range of the present invention, and the tensile strength, yield ratio, TS×EL, and |Δdσ/dε (σin)|W were within the specified ranges.

The invention claimed is:

1. A steel sheet comprising, as a chemical composition, by mass %:

C: 0.040% to 0.400%;
Si: 0.01% to 2.50%;
Mn: 0.10% to 4.00%;
Al: 0.010% to 1.500%;
P: 0.001% to 0.100%;
S: 0.0005% to 0.0100%;
N: 0.0005% to 0.0100%;
Ti: 0% to 0.200%;
Mo: 0% to 0.300%;
Nb: 0% to 0.200%;
Cr: 0% to 4.00%;
B: 0% to 0.0050%;
V: 0% to 0.300%;
Ni: 0% to 4.00%;
Cu: 0% to 4.00%;
W: 0% to 2.00%;
Ca: 0% to 0.0100%;
Ce: 0% to 0.0100%;
Mg: 0% to 0.0100%;
Zr: 0% to 0.0100%;
La: 0% to 0.0100%;
REM other than Ce and La: 0% to 0.0100%;
Sn: 0% to 1.000%;
Sb: 0% to 0.200%; and
a remainder: Fe and impurities,

wherein a microstructure in a range from a ⅛ thickness position in a sheet thickness direction from a surface of the steel sheet to a ⅜ thickness position in the sheet thickness direction from the surface includes, by area fraction,

ferrite: 10% to 97%,

residual austenite and martensite: 3% to 90%,

bainite: 0% to 87%, and

pearlite: 0% to 10%,

in a plane parallel to a rolling direction at a ¼ thickness position in the sheet thickness direction from the surface, when a maximum value of Mn contents in a measurement range is indicated as Mn_{max}, an average value of the Mn contents is indicated as Mn_{ave}, regions where the Mn content is $(Mn_{ave}+Mn_{max})/2$ or more are indicated as high Mn regions, and the other regions are indicated as low Mn regions,

an average distance between centers of the high Mn regions adjacent to each other is 1.00 mm or less,

a density D_A of the high Mn regions at a sheet width center portion and a density D_B of the high Mn regions at a ¼ width position from a sheet width end portion satisfy Expression (1),

a ratio of an average hardness of the high Mn regions to an average hardness of the low Mn regions is 1.1 to 2.0, and

a difference between an average of a top 5% and an average of a bottom 5% of the Mn contents at measurement points in the low Mn regions is 0.3 mass % or more,

$$0.77 \leq D_A/D_B \leq 1.30 \quad \text{Expression (1).}$$

2. The steel sheet according to claim **1**, wherein a hot-dip galvanized layer is formed on the surface.

3. The steel sheet according to claim **2**, wherein the hot-dip galvanized layer is a hot-dip galvannealed layer.

4. A method for producing the steel sheet according to claim **1**, comprising:

a casting step of producing a slab by melting a steel having the chemical composition according to claim **1**, casting the melted steel to produce a slab, and cooling the slab at a temperature of 950° C. to 550° C. while applying a pressure of 10 N/cm² or more to the slab in a thickness direction so that an average cooling rate is 100° C./h or faster;

a heating step of heating the slab to a temperature range of 1100° C. to 1280° C. after cooling the slab to room temperature or before cooling the slab to room temperature so that an average heating rate in a temperature range of 650° C. to 850° C. is 50° C./min or slower;

a hot rolling step of hot-rolling the slab after the heating step in a temperature range of 1050° C. or higher at a cumulative rolling reduction of 35% or more to obtain a hot-rolled steel sheet;

a cooling step of cooling the hot-rolled steel sheet to 650° C. or lower, the cooling being started within three seconds after completion of the hot rolling step, so that an average cooling rate to 700° C. is 20° C./s or faster;

a coiling step of coiling the hot-rolled steel sheet after the cooling step in a temperature range of 300° C. to 650° C.;

a pickling step of performing pickling on the hot-rolled steel sheet after the coiling step to obtain a pickled steel sheet;

a cold rolling step of performing cold rolling on the pickled steel sheet to obtain a cold-rolled steel sheet;

an annealing step of heating the cold-rolled steel sheet to an annealing temperature of Ac1° C. to 1000° C. at an average heating rate of 10.0° C./s or slower and performing holding at the annealing temperature for five seconds to 600 seconds;

a post-annealing cooling step of cooling the cold-rolled steel sheet after the annealing step to a retention temperature of 150° C. to 550° C. at an average cooling rate of 1° C./s to 200° C./s;

a retaining step of performing retention at the retention temperature for 15 seconds to 1000 seconds; and

a final cooling step of cooling the cold-rolled steel sheet after the retaining step to room temperature.

5. The method for producing the steel sheet according to claim **4**, further comprising:

a hot-dip galvanizing step of immersing the cold-rolled steel sheet in a molten zinc bath, between the retaining step and the final cooling step.

6. The method for producing the steel sheet according to claim **5**, further comprising:

an alloying step of reheating the cold-rolled steel sheet to 470° C. to 550° C. and performing holding for 60 seconds or shorter, between the hot-dip galvanizing step and the final cooling step.

7. The method for producing the steel sheet according to claim **6**, further comprising:

a leveling step of working the cold-rolled steel sheet using a leveler, between the cold rolling step and the annealing step.

8. The method for producing the steel sheet according to claim **5**, further comprising:

a leveling step of working the cold-rolled steel sheet using a leveler, between the cold rolling step and the annealing step.

9. The method for producing the steel sheet according to claim **4**, further comprising:

a leveling step of working the cold-rolled steel sheet using a leveler, between the cold rolling step and the annealing step.

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